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DEVELOPMENT OF HIGH TEMPERATURE NICKEL-BASE ALLOYS FOR JET ENGINE TURBINE BUCKET APPLICATIONS

By
H. E. COLLINS

Prepared for
NATIONAL AERONAUTICS AND SPACE ADMINISTRATION
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SUMMARY REPORT

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NATIONAL AERONAUTICS AND SPACE ADMINISTRATION

June 20, 1967

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FOREWORD

This Summary Report covers the work performed under NASA Contract NAS 3-7267 during the period from 1 April 1965 to 31 May 1967. This report has been given the NASA number CR-54507 and the TRW internal number ER-7162.

This contract was initiated between NASA Lewis Research Center and TRW Inc. for the "Development of High Temperature Nickel-Base Alloys by Conventional Alloying Techniques for Application as Jet Engine Turbine Buckets." Technical direction was supplied by the Project Manager, Mr. F. H. Harf, of the NASA Lewis Research Center, Air Breathing Engine Division, Cleveland, Ohio. Mr. R. L. Dreshfield served as the NASA Research Advisor.

Dr. H. E. Collins of TRW Equipment Laboratories was the Project Engineer. Dr. R. J. Quigg was the TRW Program Manager. Technical assistance was provided by Mr. S. T. Scheirer. Mr. A. F. Bond of TRW contributed significantly to the statistical aspects of the program.

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ABSTRACT

The object of this program was to develop a nickel-base alloy suitable for use in advanced turbine blade applications. The program was divided into two tasks.

Task I was a screening study where 75 experimental compositions were melted, cast, and evaluated on the basis of mechanical property, microstructure, and workability results. Latin Square and fractional factorial statistical designs were used to formulate the alloy compositions. The elements studied as alloy additions were Ta, W, Mo, Ti, Cb, V, Hf, Re, Ru, Al, Cr, Co, C, and Zr. Ta, W, Hf, Al, and Cr appeared to be the elements most influential in improving stress rupture life. In general, the elements and levels which increased the life also increased tensile strength, but they normally decreased ductility and workability.

In Task II, a more complete property evaluation was performed on the three most promising cast alloys and the five most promising wrought alloys derived in Task I. The results of this evaluation showed that Alloy VI A had the best combination of high temperature properties of the three cast alloys. It represents a substantial improvement in stress rupture life with an approximate 50°F increase in temperature over the lives of present-day high strength nickel-base alloys. The results also showed that four of the five alloys, Alloys I-5, IIb, IId, and IIIg, were promising wrought alloys.

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SUMMARY

The object of this program was to develop a nickel-base alloy suitable for use in advanced turbine blade applications. The program was divided into two tasks. Task I was a screening study where 75 experimental compositions were melted, cast, and evaluated on the basis of mechanical property, microstructure, and workability results. From this study, compositional ranges of alloys leading to the selection of three promising cast and five promising wrought alloys were established for more complete property evaluation in Task II.

In Task I, two statistical techniques, Latin Square and fractional factorial designs, were employed to study the effects of a number of elements on the mechanical properties and workability of the alloys. The elements studied were Ta, W, Mo, Ti, Cb, V, Hf, Re, Ru, Al, Cr, Co, C, and Zr. The level or range of levels of these elements found to produce optimum stress rupture lives with adequate ductility are as follows:

Ta	W	Mo	Ti	Cb	V	Hf
8.0-9.0	5.5-6.0	2.0	1.0	0.5-1.0	0-0.5	0.43-2.0

Re	Ru	Al	Cr	Co	C	Zr
0-1.0	0	5.4	5.4-6.1	5.0-7.5	0.13-0.15	0.03-0.13

Ta, W, Hf, Al, and Cr appeared to be the elements most influential in improving stress rupture life. In general, the elements and levels which increased life also increased tensile strength, but they normally decreased ductility and workability.

In Task II, a more complete property evaluation was performed on the three most promising cast alloys and the five most promising wrought alloys derived in Task I. The property tests for the cast alloys included stress rupture tests at 1400°F, 1875°F, and 1950°F; tensile tests in the "as cast" condition and after a 300 hour argon age at 1875°F at room temperature, 1200°F, 1400°F, 1600°F, 1875°F, and 1950°F; Charpy impact tests; thermal fatigue tests; oxidation and hot corrosion tests; and phase analysis and microstructural instability tests. The results of these tests showed that Alloy VI A had the best high temperature properties of the three cast alloys. It represents a substantial improvement in stress rupture life with an approximate 50°F increase in temperature over the lives of present-day high strength nickel-base alloys. Its tensile strength and impact properties are comparable with such alloys. It has adequate ductility, good thermal fatigue and corrosion resistance, and does not appear to suffer from microstructural instability. Its major disadvantage is cost due to the presence of rhenium.

The property tests for the wrought alloys included short and long time stress rupture tests at 1875°F. The results of these tests showed that four of the five alloys, Alloys I-5, IIb, IId, and IIIg, were promising wrought alloys.

I. INTRODUCTION

The desire for faster and larger aircraft has given impetus to the improvement of existing materials as well as the development of new materials capable of maintaining their strength and their creep, oxidation, and thermal fatigue resistance at high temperatures. The air breathing engine is a notable example where material limitations have retarded progress. The efficiency and power output of such engines are primarily dependent upon the operating temperatures. The temperatures, in turn, are dependent upon the temperature capabilities of the materials in the hot sections of the engine. As a result, improving the temperature capabilities of materials should result in greatly enhancing the efficiency and power output of the air breathing engine.

This program was initiated to develop a material with improved high temperature capabilities for use as turbine buckets. The target properties for the material are a 3000 hour stress rupture life at 1875°F under a stress of 15,000 psi. The alloy should also have resistance to cracking during thermal cycling from 1875°F to ambient temperatures. These requirements eliminated both the iron- and cobalt-base superalloy systems from further consideration. In iron-base superalloys, the loss of strength and severe oxidation become prohibitive above 1500°F[1]; while in cobalt-base superalloys, it is not currently possible to strengthen them to near the desired load bearing capacity[2]. This leaves the nickel-base superalloy system as the likely candidate for development to meet the target properties. Figure 1 shows the stress rupture properties of existent cast nickel-base superalloys in relation to the target alloy[3]. The figure shows that the target goal is a considerable increase over properties available in present-day cast superalloys.

The program was divided into two tasks. Task I was a screening study where 75 experimental alloys were melted, cast, and evaluated to determine mechanical properties, workability, and microstructure. The alloys were divided into groups of compositions formulated by statistical designs. The statistical experiments were designed to maximize the effect of the three basic alloy strengthening mechanisms available in the nickel-base alloy system: intermetallic formation, solid solution strengthening, and carbide strengthening. Task I was completed in the first fourteen months of this twenty-six month program.

Task II entailed a more complete property evaluation of the three most promising cast alloys and the five most promising wrought alloys derived in Task I. The selection of the three cast alloys was based upon the mechanical property tests performed, which included room temperature, 1400°F, 1875°F, and 2000°F tensile tests and 2000°F, 15,000 psi stress rupture tests. Stress rupture life was given the greatest weight in the selection. The selection of the five wrought alloys was based upon the workability rating and the stress rupture life. Task II was completed in the last twelve month of the program.

This report summarizes the results obtained during the twenty-six months of NASA Contract NAS 3-7267. It includes a discussion of nickel-base alloys, a review of the program to develop a high temperature nickel-base alloy, a review of the experimental procedures, a summary of the experimental results, and a discussion of these results.

II. PHYSICAL METALLURGY OF NICKEL-BASE ALLOYS

Historically, the 80 Ni-20 Cr Nimonic 75 alloy developed in England in the 1930's is considered to be the forerunner of the present day nickel-base superalloys. Since then, the addition of alloying elements has caused a gradual improvement in the high temperature strength of nickel-base alloys. This has resulted in a group of complex alloys containing a large number of such elements. Most of these alloying elements contribute to one or more of the three basic alloy strengthening mechanisms which manifest themselves in nickel-base alloy systems - intermetallic, solid solution, and carbide strengthening. Since the physical metallurgy of nickel-base alloys varies only in degree between wrought and cast alloys, a single treatment of these concepts is sufficient for both. However, there are certain aspects of wrought and cast alloy which bear separate mention.

A. Intermetallic Formation

The high temperature properties of present day nickel-base alloys depends principally upon the precipitation of intermetallic phases. Factors which influence the effects of these phases on the high temperature properties include the type, the rate of growth, the size and morphology, the density and distribution, the lattice match with the matrix, and the strength and ductility of the precipitate(s)[4,5].

In nickel-base alloys the most important single factor contributing to the retention of high temperature strength is commonly held to be the precipitation of the Ni_3Al intermetallic phase, gamma-prime. The Ni_3Al compound, like the solid solution matrix gamma, has a face-centered cubic crystal structure. While the lattice parameter of the matrix changes slightly with small changes in composition, the difference in lattice parameters between matrix and gamma-prime is very slight, i.e., generally less than 0.5 percent[6]. As a result, it can be assumed that the major strengthening effect of this precipitate, even though it is probably coherent with the matrix, is not due to lattice straining but to a variation in flow characteristics between the two phases. In fact, it has been demonstrated that the flow stress of the gamma-prime phase increases with temperature up to 1400°F[7]. Thus, the precipitated phase actually becomes more resistant to deformation, while the matrix loses strength.

A substantial portion of the aluminum atoms (up to at least 60-65%) in the Ni_3Al compound can be replaced by titanium or columbium atoms without changing the basic crystal structure or strengthening effect of the intermetallic. If this solubility limit is exceeded, orthorhombic Ni_3Cb (epsilon) or hexagonal Ni_3Ti (eta) phases will precipitate in place of or in combination with gamma-prime. Other elements may also have subordinate effects upon the gamma-prime phase. In particular, the heavy metals molybdenum and tungsten retard precipitate growth by reducing diffusion rates; cobalt is reputed to enhance the high temperature stability of gamma-prime by increasing its solutioning temperature [6].

A number of morphological forms may be associated with gamma-prime, depending both upon its stoichiometry and the thermal and mechanical history of the material. Some of these are shown in Figures 2, 3, and 4. One form, prevalent in lightly alloyed systems, is a fine intragranular precipitate which is developed by use of a solutioning and aging treatment, Figure 2A. In more heavily alloyed superalloys, it is often impossible to suppress the precipitation of gamma-prime on cooling from a solution temperature, Figure 2B. Cast alloys often have a gamma-prime dispersion favorable to high temperature properties, which is formed directly upon solidification and cooling. Such alloys, generally containing high percentages of aluminum and titanium, exhibit a coarse, somewhat blocky type of intragranular gamma-prime, Figure 3A. Heat treated alloys containing larger amounts of addition elements also exhibit a blocky intragranular gamma-prime precipitate morphology, Figure 3B. The relative coarseness of the particle depends both upon the chemistry and heat treating temperature. These forms of gamma-prime are considered to be beneficial to high temperature strength and are encouraged through the use of controlled casting techniques and selected heat treating procedures.

Another gamma-prime morphology often observed in heavily alloyed cast superalloys is the blossom-like eutectic formation of gamma and gamma-prime, Figure 4. Formed directly upon solidification, this structure can be detrimental to high temperature properties. Careful control of casting procedures in an alloy of proper composition can usually minimize this effect.

While coarser gamma-prime is relatively stable at normal service exposures, fine gamma-prime is often dissolved during extended thermal exposure. Medium size gamma-prime will also coalesce or even solution during service. Either effect may alter properties; hence, gamma-prime stability and morphology are important for effective utilization of nickel-base alloys.

B. Solid Solution Strengthening

Reference to binary phase diagrams[8] reveals that a number of elements have a wide solubility in nickel. Noteworthy among these are the refractory metal elements and cobalt. The solution of such elements in nickel increases the resistance to deformation, particularly at high temperatures. Recent trends in superalloy development have shown that increased amounts of refractory metal solid solution hardeners significantly improve the stress rupture properties, especially at 1800°F and above. Unfortunately, while increasing strength, some of the refractory metals may decrease oxidation resistance, ductility, and particularly in the case of tungsten, increase the density of the alloy.

Other elements which are taken into the nickel-rich gamma solid solution are chromium and cobalt. While not particularly beneficial to high temperature strength, these elements, nevertheless, do have important effects: chromium on oxidation resistance and carbide formation, and cobalt upon gamma-prime stability, workability, and ductility.

C. Carbides

Though the carbon level in most nickel-base alloys is quite low, e.g., in the vicinity of 0.10 percent by weight, the formation of a variety of carbides is an important contribution to the strengthening of these metals. While nickel itself is not a strong carbide former, many of the alloying constituents of superalloys do have a strong propensity to form one or more carbide types. Minor alterations in chemistry appear to exert a large influence upon the type, stability, and morphology of the carbides formed.

The most stable carbide formed in nickel-base alloys is the MC type. This carbide is generally titanium-rich, although columbium or tantalum-rich carbides will form if these elements are present. The order of decreasing preference for MC formation observed in nickel-base alloys is the following: TaC, CbC, ZrC, TiC, VC[9].

MC is generally a stable carbide close to the fusion temperature and assumes a massive, somewhat cubic structure evenly distributed throughout the microstructure, Figure 5A. The massive MC carbides do not contribute greatly to alloy strengthening. In certain instances, MC is formed in grain boundary areas, Figure 5B, where it is relatively unstable at temperatures well below the melting point.

Another common type of carbide is the $M_{23}C_6$ type. The primary metallic constituent is chromium with small amounts of molybdenum, tungsten, or iron whenever these elements are present. A typical composition is $Cr_{21}(Mo,W)_2C_6$. $M_{23}C_6$ forms at relatively low temperatures (1400°F-1700°F) and tends to re-solution at any temperature above 1800°F. It is, therefore, an effective strengthener only at lower exposure temperatures. Generally formed in the grain boundary areas, this carbide may assume one of two morphologies. If discrete particles of the carbide are formed, Figure 6A, creep resistance and ductility are enhanced, especially in moderate temperature ranges. However, a continuous carbide film or platelet may be formed in the grain boundary regions, Figure 6B. This network has an embrittling nature and a deleterious effect upon both creep resistance and ductility. The morphology of grain boundary $M_{23}C_6$ is controlled through composition, processing, and heat treatment.

Generally, the least common of the three major carbides detected in nickel-base alloys is the M_6C type. This carbide generally replaces $M_{23}C_6$ in those alloys containing over approximately 6 weight percent molybdenum plus its equivalent in tungsten. M_6C is generally a double carbide more correctly designated as M_xM_yC, where $x + y = 6$. M_x is generally one of the transition metals (Fe, Ni, or Co) and M_y is one of the refractory metals (Mo, W, or Cb). A typical composition is $NiCo_2(Mo,W)_3C$. Alloy chemistry also greatly affects the amount of this carbide formed. For example, in Inco 713C, this carbide is found in only small quantities and is relatively inert over a range of temperatures; but in an alloy such as Rene' 41 with higher molybdenum, M_6C is of major significance. Generally, it is more stable than $M_{23}C_6$, but somewhat less so than MC. M_6C assumes a discrete particle morphology, often in grain boundary regions.

Carbide phases, particularly $M_{23}C_6$ and M_6C will tend to coalesce and resolution upon extended exposure in the 1500-1900°F temperature range. These phases may then reprecipitate in a platelet or other less desirable form. It is, therefore, desirable to attain as stable a carbide formation as possible, while still retaining its creep strengthening effect. Even MC, considered the highly stable carbide, has been reported as being metastable in nickel-base alloys.

D. Minor Elements

Minor elements other than carbon which are added to nickel-base alloys include boron and zirconium. These elements were originally present as tramp elements, but attempts to lower their concentration or eliminate them completely has, in most cases, led to a deterioration in properties. Hence, most present-day alloys contain small amounts of both.

The exact function of these elements is not completely understood. It is thought that their main effect occurs in grain boundary regions. One possible explanation for their effect in these areas is that the boron and zirconium atoms migrate to grain boundary regions and fill open spaces present there. This would then slow down grain boundary diffusional processes which contribute to creep deformation and would thereby strengthen these regions(6). A second explanation supported by experimental data is that the presence of the boron and zirconium in the grain boundary retards early agglomeration of $M_{23}C_6$ carbides and the subsequent formation of creep cracks in these areas. This retardation could be due to the reduction of carbon, titanium, and aluminum segregation in the grain boundaries by boron and zirconium. This is supported experimentally by the fact that boron increases the amount of intragranular carbide precipitation which is consistent with a reduction in the grain boundary segregation of carbon(10,11). In addition, there is some experimental evidence which indicates that boron and zirconium may change interfacial energy relationships sufficiently to favor coalescence and spheroidization of secondary phases along grain boundaries(11,12). This is thought to increase grain boundary strength.

E. Wrought Alloys

From the development of the early Nimonic alloys on through the introduction of Udimet 700 in 1959, nickel-base superalloys were either wrought or were combination wrought-cast alloys. The uniformity of properties and structure, reliability, and lower cost of production of high temperature products forged from wrought nickel-base alloys were unquestioned. In fact, these alloys were used in the as-cast form only when the part could not be produced by conventional forging techniques.

The chemical composition of a few common wrought nickel-base alloys are shown, Table 1, and are designated by "W". The increasing amount of solid solution and intermetallic forming alloying elements present as one progresses from Nimonic 75 to Udimet 700 is apparent. The improvement in strength with increased alloy additions is shown in Table 2, which gives the 1800°F and 1900°F stress rupture properties. Alloys such as Nimonic 75, Inconel X, and Waspaloy are not used at temperatures as high at 1800°F. To achieve the necessary strength levels, wrought superalloys require complex heat treatments consisting of solution annealing and often several aging cycles.

In addition to the high temperature strength requirement, wrought alloys must also possess the ability to be deformed at some temperature generally above the potential use temperature. Demands for increased strength were fulfilled by increasing the amounts of alloying constituents. While additional alloying provided the required strength, it also rendered nickel-base alloys stronger at the potential deformation temperatures. As a result, wrought alloys became more difficult to produce as bar stock and forge to finished parts.

F. Cast Alloys

The advantageous position enjoyed by wrought superalloys in relation to cast alloys has today been all but nullified when strength retention above 1800°F is demanded. Casting practice has improved tremendously to the point where uniformly sound, reliable products are often the rule. The advantages of cast alloys as creep resistant materials above 1800°F are two-fold: (1) a given alloy, in the cast form is more creep resistant than in the wrought form, probably because of grain size and subgrain effects; (2) increased amounts of aluminum and titanium and the solid solution strengthening refractory elements can be added with the removal of the workability criterion.

Typical compositions for a number of currently prominent cast nickel-base superalloys are shown in Table 1. While increased amounts of alloying elements are evident when compared with the wrought alloys, the trend is also apparent from an earlier cast Inco 713C to the recently developed NASA alloys TAZ8 and TAZ8A. The improved strength derived from these alloy additions is presented in Table 2.

A further advantage of cast alloy materials is the ability to be precision cast to hollow configurations. Providing that mold technology and alloy castability is sufficient, a wide variety of intricate parts can be investment cast by the lost wax method.

III. PROGRAM FOR THE DEVELOPMENT OF A HIGH TEMPERATURE NICKEL-BASE ALLOY

The program for the development of a high temperature nickel-base alloy was divided into two tasks:

A. Task I - Screening Studies

B. Task II - Complete Property Determination

A. Task I - Screening Studies

Task I involved a screening study where 75 experimental compositions were melted, cast, and evaluated to determine mechanical properties, workability, and microstructure. The 75 compositions were divided into groups of alloys utilizing statistical designs to formulate compositions for each of the alloys in a given group. The statistical experiments were designed to maximize the effect of the three basic alloy strengthening mechanisms available in the nickel-base alloy system: intermetallic formation, solid solution strengthening, and carbide strengthening. In this section, the requirements and evaluation of the proposed alloy and the methods of statistical analysis utilized in alloy selection are presented.

1. Requirements and Evaluation of the Proposed Alloy

In order to satisfy the requirements for a turbine blade material in the new higher temperature air breathing engines, a substantial increase in the presently available properties of nickel-base superalloys is required. Guidelines, which were established as target properties for the new nickel-base material, are the attainment of 3000 hour stress rupture life in an 1875°F test under 15,000 psi load and the resistance to cracking during repeated cycling from 1875°F to ambient temperature. The stress rupture capability of several present-day cast nickel-base alloys has already been noted in Figure 1. It is apparent that considerable upgrading of present-day alloys is required to achieve an alloy with the suggested target properties.

In order to avoid extended testing times, stress rupture evaluation in Task I was carried out with a screening test at 2000°F and 15,000 psi stress level. The tests were performed in air. When tested under these conditions, lives from 100 to 150 hours can be considered to roughly correlate with the 3000 hour criterion at 1875°F for this type material.

In addition to evaluation of experimental compositions by stress rupture testing, tensile testing at room temperature, 1400°F, 1875°F, and 2000°F were carried out to ascertain further the overall capacity of the nickel-base superalloys. These tests reveal the maximum strength inherent in the alloy and the ductility it possesses. Ductility is particularly important in the 1400°F temperature range where many nickel-base superalloys are severely limited due to low ductility.

2. Statistical Analysis and Alloy Selection

In order to achieve the proposed goals of this program, it is evident that the nickel-base alloy developed must contain optimum combinations of the three basic strengthening elements: intermetallic, solid solution, and carbide strengthening elements. In Task I of this program, suitable combinations were developed by grouping elements by their alloy types and then analyzing the mechanical properties of each alloy grouping individually and collectively (including interactions where applicable) through suitable statistical techniques.

The 75 compositions were divided into six groups or series of alloys: 27 in Series I, 9 each in Series II and III, 18 in Series IV, 8 in Series V, and 4 in Series VI. Two statistical techniques, Latin Square and fractional factorial, were utilized to aid in the formulation of the alloy compositions. These two techniques will be reviewed briefly below, and the designs and/or elements varied in each series presented.

The Latin Square design[13] is particularly useful where variables can be separated into groups of three, ideally having a similar function, as is often the case in nickel-base superalloys. Each of the three elements in a given grouping are assigned three compositional levels for a total of nine alloys for each square. By analyzing the sum of squares and using the "F" test for significance, the individual consequence of each element can be ascertained and the total interaction between the elements can be estimated. One is then able to discern the level of confidence for certain data trends. By varying an element over a given range using three values, it can be determined whether or not an element has a significant effect upon the property changes which develop as the chemistry is altered.

The Latin Square design was used in Series I, II, and III. Five Latin Squares were set up - three in Series I and one each in Series II and III. These are shown in Table 3. The elements held constant for all Series I, II, and III compositions (total 45 alloys) and the levels employed are given below (weight percent):

Co	Cr	Zr	B	C
10.0	10.0	0.03	0.02	0.13

The principal purpose of Series I was to optimize the solid solution strengthening elements Mo, W, and Ta. Previously developed nickel-base compositions, such as TRW 1900, SM 200, and B 1900, have shown that these elements in some quantity can produce significant strengthening. However, since the amount of these elements can greatly affect the amount of nickel available for gamma-prime formation, the relative amounts of these elements must be considered in conjunction with three Latin Square designs of nine compositions each. Mo, W, and Ta were varied at the same three levels in each square with each square consisting of different (Al + Ti) levels, as shown in Table 3 (Squares 1, 2, and 3). As a result of the analysis of Series I data, the following fixed additions were designated for Series II and III:

Mo	W	Ta	Ti
2.0	5.5	8.0	1.0

The object of Series II was to determine the effect of small additions of the strong carbide formers, Cb, V, and Hf. Columbium and vanadium have been found to be particularly important for property optimization in present-day alloys; e.g., Cb in TRW 1900 and SM 200 and V in IN 100 and TAZ-8. Hafnium should have similar effects. Cb, V, and Hf were varied in Square 4 (Table 3) using the elemental levels designated previously and 4.5 Al.

In Series III, the purpose was two-fold. The first was to determine the effect of two elements, Re and Ru, not commonly used in nickel-base alloys. These elements show some solid solubility plus sufficient lattice mismatch to provide potential strengthening and have been advocated in another similar alloy development effort[14]. The second was to optimize the strengthening effect of Al, the principal gamma-prime former. Re, Ru, and Al were varied in Square 5 (Table 3) using the new base.

It became evident after reviewing the results of the first three series that further testing was required to optimize the levels of several of the elements being varied, such as Ta, W, Hf, and Re. In addition, several importand elements appeared to require further study. These included two major additions, Cr and Co, and two potent minor additions, C, and Zr. To study this number of variables using the Latin Square design would have required a minimum of three squares of 9 alloys each or 27 out of the 30 remaining alloys. This would have allowed only three alloys for further alloy optimization. Fractional factorial design, on the other hand, is particularly well suited for a screening study on large numbers of variables with a reasonably small number of experiments. The design utilized is orthogonal and permits the calculation of main effects for each component incorporated into the design[15]. The main effect of an element for a given property parameter is defined as the difference between the average value of that property for alloys with the element at the high concentration level and that with the element at the low concentration level. Main effects are either positive or negative with a positive value indicating improvement in the property with increasing concentration of the element. However, one major disadvantage of the two level factorial over the Latin Square technique is the possibility that the property goes through a minimum (or maximum) between the two levels. This design was used in Series IV and V.

The primary objective of Series IV was to determine the effects of Cr, Co, C, and Zr and to further examine the effects of Ta, W, Hf, and Re on the mechanical properties. A fractional factorial experiment consisting of 16 alloys was set up in which 8 factors (the 8 chemical elements) were varied at 2 levels. Plan 8 of Table 12-4 of Reference 16 was modified to obtain the desired experimental design. This factorial plan was modified so that several alloys out of the 16 contained the high percentages of only two of the eight elements being varied instead of four as in Plan 8. This was accomplished by formally multiplying each Plan 8 treatment combination by the letters gh and deleting g^2 and h^2 wherever they occur. Both plans possess the property that the eight main effects are obtained separately and not confounded with first order interactions, and that the 28 first order interactions are obtained in seven separate confounded groups of four (each one confounded with three others) in such a way that each group of four involves all eight factors. These two plans, the levels of the elements being varied, the alloy designations, and the design symbols are given in Table 4. The elements held constant in Series IV are given below:

Al	Mo	Ti	Cb	B
5.4	2.0	1.0	0.5	0.02

In addition to the 16 alloys noted above, an alloy based upon a combination of past experience in the program and metallurgical judgment and an alloy with the design-center composition of the fractional factorial experiment (Alloys Y and Z, respectively) were included in this series, bringing the number of alloys in Series IV to 18. The nominal compositions of these two alloys are given below:

Alloy	Co	Cr	Ta	W	Hf	Re	Zr	C	Al	Mo	Ti	Cb	B	Ni
Y	5.0	6.0	8.0	5.5	2.0	1.0	0.03	0.15	5.4	2.0	1.0	1.0	0.02	Bal.
Z	7.5	8.5	9.0	6.5	2.0	0.5	0.13	0.13	5.4	2.0	1.0	0.5	0.02	Bal.

Series V was designed to provide the information required so that it, in conjunction with Series IV, could be used to determine the constants of a general quadratic equation which would then be used for final optimization of four components. A fractional factorial experiment consisting of eight alloys (out of 12 remaining alloys for Task I) was set up. These alloys formed a fractional factorial experiment in which four chemical components were varied at two levels. Plan 2 of Table 12-4 of Reference 16 was the design used. The four chemical components varied - Cr, Ta, W, and Hf - were selected from Series IV as being the four (out of eight) components having the most significant effect on the mechanical properties of the alloys, especially stress rupture life. Plan 2, the levels of the elements being varied, the alloy designations, and the design symbols are given in Table 5. The elements held constant in Series V are given below:

C	Mo	Ti	Al	Co	Re	Zr	B	Cb
0.13	2.0	1.0	5.4	7.5	0.5	0.13	0.02	0.5

Series VI consisted of four alloys, thereby completing the 75 alloys of Task I. One of the four alloys (Alloy VI A) was determined from the aforementioned solution of the general quadratic equation of four components (Cr, Ta, W, and Hf) for maximum stress rupture life. The remaining three alloys were determined from a combination of experience gained from the prior alloy series of this program and metallurgical judgment. The nominal compositions of these four alloys are given below:

Alloy	Cr	Ta	W	Hf	C	Mo	Ti	Al	Co	Re	Zr	B	Cb	V
A	6.1	9.0	5.8	0.43	0.13	2.0	1.0	5.4	7.5	0.5	0.13	0.02	0.5	0
B	6.0	8.5	6.0	1.75	0.15	3.0	1.0	5.4	5.0	0.5	0.08	0.02	1.0	0.5
C	6.0	8.0	6.0	1.75	0.15	2.0	1.3	5.4	5.0	0.5	0.08	0.02	0.75	0
D	5.4	8.5	6.0	1.75	0.15	2.0	1.0	5.4	5.0	0	0.08	0.02	0.5	0.5

B. Task II - Complete Property Determination

Task II entailed a more complete property evaluation of the three most promising cast alloy candidates and the five most promising wrought alloy candidates from Task I. The alloy candidates were selected in the following manner: the cast alloys selection was based upon the mechanical properties with emphasis on stress rupture life; and the wrought alloys, based upon the workability results and stress rupture life. The nominal compositions and average 2000°F/15,000 psi stress rupture lives of the three cast alloys are given in Table 6. The nominal compositions, average 2000°F/15,000 psi stress rupture lives, and workability ratings of the five wrought alloys are given in Table 7. In this section, the property evaluations performed in Task II on the most promising cast alloys and on the most promising wrought alloys are outlined.

1. Evaluation of Cast Alloys

The following property determinations were made on each of the three cast compositions:

a. Tensile tests were run in air on the "as-cast" test specimens at room temperature, 1200°F, 1400°F, 1600°F, 1875°F, and 1950°F.

b. "As-cast" test specimens were aged in an argon blanket for 300 hours at 1875°F. Tensile tests were then run at room temperature, 1200°F, 1400°F, 1600°F, and 1875°F.

c. Stress rupture tests were run on "as-cast" test specimens in air until failure. Creep curves were also developed for each alloy under each test condition. The test conditions were as follows:

(1) 1400°F under 85,000, 90,000, and 94,000 psi load.
(2) 1875°F under 15,000, 25,000, and 35,000 psi load.
(3) 1950°F under 15,000, 20,000, and 25,000 psi load.

d. Stress rupture tests were run on "as-cast" test specimens in commercially pure argon at 1875°F and 1950°F under a 15,000 psi load.

e. Oxidation tests were run on "as-cast" test specimens in air for 50, 200, 500, and 1000 hours at 1875°F, 1950°F, and 2100°F. Also, a hot corrosion crucible test was run on these specimens at 1800°F.

f. Charpy unnotched impact tests were run on "as-cast" specimens in air at room temperature, 1400°F, and 1875°F.

g. Thermal cycling resistance tests were run in air on "as-cast" specimens from 1875°F and from 2100°F to ambient temperature.

In addition to the above, the following techniques for structural evaluation were employed as necessary:

h. Photomicrographs were made showing the general appearance of the fracture of a typical stress rupture bar for each test condition.

i. Light or electron microscopy was used to evaluate the "as-cast" structure, the "as-heat-treated" structure, and the mode of fracture in a typical stress rupture bar for each test condition.

j. Phase analyses were conducted on several selected alloy conditions to determine the carbides present. In addition, after a 1500 hour age at 1600°F, phase analyses were used on cast alloys to determine whether they were microstructurally stable, i.e., whether they formed detrimental phases, such as sigma, mu, or Laves.

2. Evaluation of Wrought Alloys

Stress rupture properties were determined for the five wrought compositions in the following manner:

a. Three inch diameter cast ingots were double extruded to form nominal 1/2 inch bar.

b. Solution heat treatments were developed to optimize 1875°F stress rupture properties. Standard aging cycles were employed.

c. Test specimens were then heat treated in an argon blanket using the treatment developed above.

d. Stress rupture tests at two stress levels were conducted in air. The stress levels were selected to produce rupture in a relatively short time, such as ten hours, and a relatively long time, such as three hundred hours, at 1875°F.

In addition to the property determinations, the techniques for structural evaluation noted above were employed as necessary.

IV. MATERIALS AND PROCEDURES

The experimental techniques employed in this program are those normally employed to evaluate cast nickel-base superalloys. The important aspects of these procedures for each task will be described in this section. Note that all of the procedures employed in Task I except for workability testing were also employed in Task II. As a result, only changes or additions to these procedures will be described for Task II.

A. Task I

1. Materials

In order to minimize the effects of impurity elements, raw materials of the highest purity which could be economically obtained were employed. A purity level of 99.9 percent was specified on most alloying elements. Alloying elements with an impurity content higher than 0.1% are listed in Table 8. In all cases a certified chemical analysis of the raw material was obtained from the supplier.

2. Melting and Casting Procedures

A two step melting and casting procedure was employed for each heat. In Step 1, the virgin alloying elements were melted in a NRC vacuum induction furnace using a Magnorite crucible. The virgin metal was cast as a cylindrical ingot. The melting procedure utilized in Step 1 is outlined in Table 9. One heat was made for each experimental alloy. The weight of charged material per heat was 20 pounds for Series I alloys and 24 pounds for Series II to VI alloys. Of this weight, approximately 2.5 to 3 pounds was cast into a patty for chemical analysis while the rest of the material was cast into the master ingot. The freezing point of each heat was determined by immersing a Pt-Pt 13% Rh thermocouple protected by a silica protection tube into the melt and recording the temperature at which small nuclei, termed "silverfish", began to collect in the liquid bath. Wash heats of pure nickel were employed whenever it was necessary to sinter in new crucibles or to melt an alloy with zero concentration of an element which was present in the previous alloy. The wash heats served to scavenge all potential contaminants.

In Step 2, the virgin metal was remelted in either a NRC or a Stokes vacuum induction furnace and cast into a tensile bar cluster. A tensile bar cluster yielded fourteen tapered 0.250 inch gauge diameter tensile bar preforms (minimum diameter approximately 0.275 inch). In addition, each tensile bar cluster yielded two square bars, 1/2" x 1/2" x 3", and a 1 inch diameter "down pole" about 3 inches in length. The melting procedure utilized for each heat in Step 2 is outlined in Table 10. All remelting was conducted in disposable ceramic liners, thereby eliminating the need for wash heats. The molds were imbedded in ceramic and preheated to 1700°F. The ceramic prevented excessive loss of heat by the preheated molds while the metal was being remelted. The preheated molds retarded the solidification rate so that sound, shrink-free castings were obtained with little segregation.

3. Chemical Analysis

A complete chemical analysis for all alloying elements and important impurities, such as Fe, Mn, Si, and S, was performed for each heat. The analyses were performed using a combination of X-ray spectrographic, spectroscopic, and wet techniques depending on the element analyzed. In a number of cases, the X-ray spectrographic and spectroscopic results were compared with the wet analysis results in order to check out the standards which had been set up for these alloys.

4. Inspection

Following the cutting of the test bars from the cast clusters, all bars were radiographically inspected. Each specimen was radiographed in one position, rotated 90°, and radiographed again to be sure that no defect would go undetected because of orientation. This technique is very useful in detecting porosity, cracks, or inclusions resulting from the casting process. After machining to size and threading, each specimen was inspected for surface cracks using the fluorescent dye penetrant technique.

5. Mechanical Testing

Two types of mechanical tests were employed as screening tests for the 75 experimental alloys. The tests consisted of stress rupture testing and tensile testing of "as-cast" material in air.

(a) Stress Rupture Testing

Stress rupture testing was performed in air at 2000°F under 15,000 psi load. The tests were conducted on Satec constant load, lever arm machines according to ASTM standards. Temperature control was maintained to within ±3°F. Triplicate specimens were run for all stress rupture tests.

(b) Tensile Testing

Tensile testing was performed in air at room temperature, 1400°F, 1875°F, and 2000°F. The tensile tests were run on an Instron Universal testing machine at a constant crosshead speed of 0.020 inch per minute. One exception occurred in Series I where the majority of the room temperature and 1400°F tests were run on a Baldwin Testing Machine. A strain rate of 0.005 inch per inch per minute as measured with a strain pacer was used with this machine. ASTM standards and furnace control to within ±3°F were adhered to. Duplicate specimens were run for all tensile test conditions.

6. Metallography

The specimens were mechanically polished and etched with the chemical etching reagent: 62% H_2O, 15% H_2SO_4, 15% HF, and 8% HNO_3. A transverse section from the thread portion of a room temperature tensile bar of each alloy was used for metallographic examination of the "as-cast" structure. Photomicrographs were made of each alloy at 250X and 750X magnification. Representative specimens were also examined by electron microscopy using the replica technique.

Photomicrographs at 5X magnification were made of representative samples of the stress rupture specimens. To examine further the mode of fracture, longitudinal sections through the necked down portion of these specimens were photographed at 60X magnification.

7. Workability Testing

The workability of the 75 experimental alloys was determined by extruding the "down pole" portion of the tensile bar cluster and side pressing up to 50 percent reduction the resulting stock. The down poles were X-rayed for porosity prior to extrusion. They were then machined to approximately 0.90 inch diameter and clad in mild steel to 1.50 inch diameter. These billets were preheated to 2075°F and extruded at a reduction ratio of 4:1 to produce nominal 1/2" bar stock. Exceptions to this procedure occurred for Alloys B, C, D, E, F, G, and H of Series V. Alloy C was extruded at 2150°F. To extrude the other six alloys, it was necessary to use a reduction ratio of 3.1:1 at a temperature of 2150°F. The extrusions were performed on a 150 ton Lake Erie Forge Press. The extruded bars were machined to sound superalloy bars approximately one half to one inch in length and 0.325 and 0.375 inch in diameter. They were then deformed by sidepressing up to 50% of the diameter between flat dies on the 150 ton Lake Erie Forge Press at temperatures from 2050°F to 2150°F. The sidepressed bars were then rated by several observers using the rating scale shown in Table 11 with the average rating being reported to the nearest half division.

B. Task II

1. Melting and Casting Procedures

In Task II, three virgin heats were made for each cast alloy, and one virgin heat for each wrought alloy. The weight of charged material per virgin heat was 50 pounds. The remelt procedure for a heat varied depending upon whether it was a cast alloy or a wrought alloy. The procedure for a cast alloy is the same as in Task I, Table 9, while that for a wrought alloy is given in Table 12.

For cast alloys, six test bar clusters were required for each alloy - four tensile bar clusters, one special cluster for Charpy bars, and one special cluster for thermal fatigue specimens. Each virgin heat yielded two test bar clusters, thereby requiring three heats for each alloy. One tensile bar cluster was cast from each heat plus either one of the special clusters or a second tensile bar cluster. A tensile bar cluster yielded sixteen tapered 0.250 inch gauge diameter tensile bar preforms (minimum diameter approximately 0.275 inch). A special cluster for Charpy bars yielded sixteen square bars, 1/2" x 1/2" x 3". A special cluster for thermal fatigue specimens yielded sixteen cast wedges, approximately 2" long, 3/8" at the large end, and 13/32" wide.

For wrought alloys, one extrusion ingot was required for each alloy. The mold for this ingot was the "flower pot" type, a sketch of which is shown in Figure 7. This mold was designed to produce the fine grain structure required for extrusion and holds approximately 25 to 30 pounds of material. It was not preheated. The ingot was removed from the mold immediately after being removed from the furnace and was then cooled in vermiculite.

2. Mechanical Testing

Four types of mechanical tests were employed in order to evaluate the mechanical properties of the alloys. The tests included stress rupture testing in air and in argon atmospheres, tensile testing of "as-cast" and "heat-treated" material, Charpy impact testing, and thermal fatigue testing.

(a) Stress Rupture Testing

Testing was performed in air and in commercially pure argon atmosphere for the cast alloys and in air for the wrought alloys under the conditions specified above in the Task II program outline. In the argon tests, the chambers were flushed with argon before heating; and then the atmosphere was maintained after the test until the specimen was cooled to less than 1000°F. Triplicate specimens were run for all test conditions.

Creep curves were developed for the cast alloys tested in air. Three different techniques were utilized to develop these curves. They are: (1) platinum extensometer with a measuring telescope, (2) mechanical extensometer with continuous electronic recording, and (3) dial gauge measurements. Techniques 1 and 2 have a sensitivity of 1×10^{-5} inches while technique 3 has a sensitivity of 1×10^{-4} inches. One curve was developed with either technique 1 or 2 for each alloy under each test condition and then a second curve with technique 3.

(b) Tensile Testing

Tensile tests were performed on "as-cast" and "heat treated" material at the temperatures specified in the Task II program outline. The heat treatment was a 300 hour age at 1875°F (furnace control ±5°F) in an argon blanket. Based upon the metallographic examination of control samples, the test bars were machined in such a way that there was no internally oxidized or alloy depleted areas in the test sections. Duplicate specimens were run for all test conditions.

(c) Charpy Impact Testing

The Charpy impact testing was performed on a Baldwin Sonntag Impact Tester in accordance with ASTM standards. The pendulum velocity was 17 ft/sec. The test specimens were unnotched with dimensions of 2.165" x 0.394" x 0.394" and were machined from "as-cast" material. Quadruplicate specimens were run for all test conditions.

(d) Thermal Fatigue Testing

The thermal fatigue resistance was determined utilizing the apparatus shown in Figure 8. This unit uses alternate impingement of oxy-acetylene flame and air cooling to produce thermal cycling within the test specimen. Both the flame and the air blast are regulated by metering controls. In the unit, only one end of the thermal fatigue specimen is anchored in the apparatus holder. Flame impingement on the specimen creates a thermal gradient, and restraint is provided by the cool areas of the sample farthest from the flame. Internal stresses are set up which cause eventual cracking of the material.

Two thermal cycles were used to evaluate the alloys - from 1875°F and from 2100°F to ambient temperature in air. The maximum test temperatures were controlled to within ± 25°F. Temperature measurements were made with an optical pyrometer focused on the hot zone of the test specimen and were made nominally at 10, 20, 50, 100, ... cycles for each crack examination interval. The time required for a complete cycle was one minute with 30 seconds for heating and 30 seconds for cooling.

The test specimens were precision cast wedges 2" long, 13/32" wide, 3/8" at the back face and 1/32" at the narrow face. The flame or air blast impinged directly on the narrow face. Duplicate specimens were run at 1875°F and quadruplicate specimens at 2100°F.

A crack was detected by careful examination of the hot zone with a 30 power binocular microscope. Examination was made nominally at 50, 100, 150, 200, 300, ..., 1000, 1200, ..., 2400, 2800, ..., 4000 cycles for the 1875°F thermal cycle and at 30, 60, ..., 150, 200, ..., 400, 500, ..., 1500, 1700, ..., 3100, 3600, ... cycles for the 2100°F thermal cycle. Occasionally in examination, the specimen surface was dampened with alcohol to delineate clearly a crack, particularly one in the initial stages of propagation. A sample was judged to have failed when the crack propagated completely across the narrow face.

3. Corrosion Testing

Oxidation and hot corrosion tests were employed to evaluate the corrosion resistance of the alloys. A gravimetric method was employed in the oxidation testing. Test specimens were prepared by machining and grinding cylinders 0.5" in diameter by 0.5" long to provide a uniform surface for testing. The specimens were placed in high purity alumina crucibles and weighed before heating. New crucibles, which had been fired for several hours at 2200°F, were used at 2100°F. After being cleaned of the spalled scale, these crucibles were reused at 1875°F and 1950°F. The same crucible was utilized for an alloy and test time at the three test temperatures. The specimens in the crucibles were heated in air at one of the test conditions. One specimen per alloy was run for each test condition. Upon removal from the furnace, the crucibles were covered to prevent loss of scale due to flaking during cooling. After cooling, the crucible and contents were weighed to determine the weight change as a result of oxidation. In addition to the weight measurements, the specimens were examined metallographically after 1000 hours to determine oxide penetration. Spalling and other surface conditions were recorded.

The TRW crucible test was used for hot corrosion (or sulfidation) testing. This is a gravimetric method which uses the same size specimen as the oxidation tests. The specimens were weighed and then tested. Specimens were partially covered with a 1% sodium chloride, 99% sodium sulfate mixture in silica crucibles. The crucibles with the specimens and salt mixture were heated in air for 1 hour at a temperature of 1800°F. The scale produced during the corrosion of the specimens was removed by cathodically descaling in molten sodium hydroxide. After descaling, the specimens were reweighed and the loss of weight used as a measure of the sulfidation attack.

4. Metallography

The metallographic examination of the alloys included both light and electron microscopy. The specimens were mechanically polished and etched with one or a combination of the following etches, depending on the response of the particular alloy:

(a) 62% H_2O, 15% H_2SO_4, 15% HF, and 8% HNO_3.

(b) 10% H_2O_2 in HCl.

(c) 15ml H_2SO_4 in 100ml CH_3OH, electrolytic etch.

A transverse section from the thread portion of a room temperature tensile bar of each alloy was used for metallographic examination of the "as-cast" structure. In addition, transverse and longitudinal sections from the thread portion of a room temperature tensile bar of each alloy were used for metallographic examination of the structure after the 300 hour age at 1875°F in argon. Photomicrographs were taken at 250X and 750X magnification. Specimens were examined by electron microscopy using the replica technique.

Photomicrographs at 5X magnification were taken of representative samples at each stress rupture test condition for each alloy. To examine further the mode of fracture, longitudinal sections through the necked down portion of the representative samples were photographed at 60X, 250X, and 750X magnification.

For the wrought alloys, photomicrographs were made at 250X and 750X magnification of longitudinal and transverse sections of materials in the "as-forged" and "as-heat-treated" conditions. Similarly, photomicrographs were made of a representative specimen in the "necked down" portion for each alloy at each stress rupture test condition.

5. Phase Analysis

Phase analysis was conducted on eight alloy conditions. They were the three cast alloys in the "as-cast" condition and after 1500 hours at 1600°F, and Alloy VI A after a 1875°F, 15,000 psi stress rupture test and after a 300 hour argon age at 1875°F. The objective was to identify the secondary phases present, e.g., carbides, borides, Laves, sigma. The procedure employed was to extract these phades electrolytically in a 10% HCl, 90% methanol solution. The digestion process varied from 12 to 24 hours at approximately 0.75 amp per square inch. The residues were separated by centrifuging, washed in methanol, and dried. The X-ray powder patterns of the residues were recorded on a diffractometer with copper K α radiation at a scanning speed of 1/2° per min. The diffraction patterns were then compared to known X-ray patterns in the ASTM card index to identify the phase(s) in the alloys.

6. Extrusion

The extrusion ingots for the wrought alloys were prepared for the primary extrusion by machining into billets approximately 2-3/4 inches in diameter by 6 inches long which were clad in 304 stainless to 3.00 inch diameter. The billets were preheated at temperatures ranging from 2025°F to 2100°F and extruded at a reduction ratio of 4:1. Primary extrusion was accomplished on a 700 ton Loewy extrusion press. The primary extruded material was then cut into approximately 3 inch lengths.

The primary extruded lengths were prepared for the secondary extrusion by machining to a billet size of one 1-1/4 inches in diameter. These were clad in mild steel to 1.50 inch diameter. These billets were preheated to 2075°F and extruded at a reduction ratio of 4:1. The secondary extrusion produced nominal 1/2 inch bar stock. Secondary extrusion was performed on a 150 ton Lake Erie Forge Press. This material was then heat-treated to optimize 1875°F stress rupture properties for each alloy and machined to 0.160 inch diameter stress rupture specimens.

V. SCREENING STUDIES - TASK I

The results obtained in Task I of the program are presented and discussed in this section. The results include the pertinent melting data, the chemical analyses, the mechanical property test data, and the workability results.

A. Results

The pertinent melting data for Series I to VI are given in Table 13. The data include melting times and "silverfish" temperatures of virgin heats and casting times and vacuum pressures at 2900°F of remelt heats.

The chemical analyses and the mechanical property test data for Series I to VI are presented in Tables 14 to 49 using the following format for each series: chemical analysis, 2000°F stress rupture data, and tensile test data at temperatures of room temperature, 1400°F, 1875°F, and 2000°F. The chemical analysis tables include the aim and actual compositions. The 2000°F stress rupture tables include life in hours, $\log_{10}$ of life in hours, percent elongation, and percent reduction in area. The tensile test tables include ultimate tensile strength, 0.2% offset yield strength, percent elongation, and percent reduction in area. Series I results are given in Tables 14 to 19; Series II results, in Tables 20 to 25; Series III results, in Tables 26 to 31; Series IV results, in Tables 32 to 37; Series V results, in Tables 38 to 43; and Series VI results, in Tables 44 to 49.

The workability results for Series I to VI are presented in Tables 50 to 55 respectively. These tables include the number of pieces pressed, the percent reduction, the temperature of reduction, and the rating for each material. The definitions of the rating scale are given in Table 11.

B. Discussion

1. Castability and Machinability

The castability of all Task I alloys compared favorably with that of other contemporary nickel-base cast alloys. Qualitative observations of the melting and casting operations indicated that the alloying ingredients were easily melted within the limits of the established procedure. Reactive metals were added without violent reactions or metal loss. Fluidity at the pouring temperature was excellent. All heats were remarkably free of impurity slags. In addition, the recovery of the alloying elements was generally quite good as is shown in Tables 14, 20, 26, 32, 38, and 44. Note that the hafnium contained from 1.5 to 2.0 Zr, making zirconium levels slightly higher than anticipated in alloys, such as Alloy IV Y.

The pertinent melting data for these alloys are given in Table 13. Variations in the time required for a heat were primarily due to differences in the melting time, pump down time, and duration of the carbon boil for the virgin heats. When the furnace was used after a period of inactivity, attaining the proper vacuum was slower. Also, when the crucible had been used previously and was thoroughly heated, melt-down occurred more rapidly.

Further verification of the castability of these alloys was the complete absence, with one exception, of porosity detectable by standard radiographic techniques in the tensile specimens. There was only one tensile bar rejected among the 75 heats for this reason. Some of the "down poles", which were subsequently extruded for workability studies contained some degree of porosity near the upper end as revealed by X-ray inspection. However, these portions were largely removed before extruding. Rejection of machined specimens in the dye penetrant test was also minimal. Only three finish ground test bars were withheld from testing for small imperfections. There was no evidence of cracking or surface checking from machining or grinding operations. Hence, it appears that all of the Task I alloys fulfill the requirement of adequate castability and also possess the necessary machinability.

2. Mechanical Property Tests

The results of the mechanical property tests for a series were used as a basis for formulating the alloy compositions of subsequent series. Heavy emphasis was placed upon the 2000°F stress rupture tests in this phase of the program since the target goal was 3000 hours stress rupture life at 1875°F under 15,000 psi load.

(a) Series I Alloys (27 Compositions)

In Series I, the effects of additions of the solid solution strengthening elements, W, Mo, and Ta, were studied using the Latin Square design. The levels of the elements used were 1.0, 4.5, and 8.0 for Mo; 1.0, 5.5, and 10.0 for W; and 1.0, 4.5, and 8.0 for Ta. The effect of these elements were evaluated at three different levels of the gamma-prime formers, (Al + Ti). The levels of (Al + Ti) used were 4.5 Al + 1.0 Ti, 6.3 Al + 1.0 Ti and 6.3 Al + 1.8 Ti (Squares 1, 2 and 3 of Table 3).

Statistical results of the 2000°F stress rupture life using logarithmic functions and of the ductility using percent elongation are shown graphically in Figures 9 and 10, respectively, where the average test values are plotted versus composition in weight percent. (The results of the Latin Square analyses are tabulated in Appendix I along with a sample calculation.) The results of this analysis show that all elements have a statistically significant positive effect on 2000°F stress rupture life in a base containing 4.5 Al + 1.0 Ti, i.e., increasing concentration from the low to the high level increases the life. In a base with 6.3 Al + 1.0 Ti, only Ta has a statistically significant effect, a positive one; and none of the refractory elements investigated have a statistically significant effect in a base with 6.3 Al + 1.8 Ti. All of the elements appear to have a statistically significant negative effect on 2000°F stress rupture ductility in a base containing 4.5 Al + 1.0 Ti. Both Mo and Ta have a significant effect in a 6.3 Al + 1.0 Ti base with Mo going through a minimum near 4.5 and Ta having a negative effect. In a 6.3 Al + 1.8 Ti base, only W has a significant effect, going through a minimum at about 5.5.

Statistical analyses of ultimate tensile strength and percent elongation at temperatures of room temperature, 1400°F, 1875°F, and 2000°F are shown graphically in Figures 11 to 18. These results indicate that the statistically significant effects which each of the elements has on these properties are as follows: Mo has a positive effect on 1400°F, 1875°F, and 2000°F tensile strengths in a 4.5 Al + 1.0 Ti base. It has a negative effect on room temperature ductility in all three (Al + Ti) bases, on 1400°F ductility in 6.3 Al + 1.0 Ti and 6.3 Al + 1.8 Ti bases, and on 1875°F ductility in a 4.5 Al + 1.0 Ti base. W has a positive effect on room temperature, 1875°F, and 2000°F tensile strengths and goes through a minimum near 5.5 for 1400°F tensile strength with a strong positive effect at the higher level of 10.0, all in a 4.5 Al + 1.0 Ti base. It has a negative effect on room temperature ductility for all three bases, on 1400°F ductility in a 6.3 Al + 1.0 Ti and 6.3 Al + 1.8 Ti bases, and on 2000°F ductility in a 4.5 Al + 1.0 Ti base. Ta has a positive effect on room temperature, 1400°F, 1875°F, and 2000°F tensile strengths in a 4.5 Al + 1.0 Ti base. It also has a positive effect on 2000°F tensile strength in a 6.3 Al + 1.0 Ti base and goes through a minimum for room temperature tensile strength near 4.5 with a strong positive effect at the higher level of 8.0. Ta has a negative effect on room temperature and 1875°F ductility in a 6.3 Al + 1.0 Ti base and on 2000°F ductility in a 4.5 Al + 1.0 Ti base, and goes through a maximum at about 4.5 for room temperature and 1400°F ductility in a 6.3 Al + 1.8 Ti base, for 1400°F ductility in a 6.3 Al + 1.0 Ti base, and for 1875°F ductility in a 4.5 Al + 1.0 Ti base.

While the primary purpose of this work was to statistically analyze the effects of given elements, certain additional benefits can be derived from the test results. Examination of the test data, Tables 15 to 19, shows that the average 2000°F stress rupture lives of three alloys stand out. They are Alloys 5 (1.0 Mo, 5.5 W, 4.5 Ta, and (6.3 Al + 1.0 Ti)), 7 (1.0 Mo, 10.0 W, 8.0 Ta, and (4.5 Al + 1.0 Ti)), and 13 (4.5 Mo, 5.5 W, 8.0 Ta, and (4.5 Al + 1.0 Ti)) with lives of 11.9, 16.4, and 8.5 hours, respectively. These alloys (Alloy 7, in particular) compare favorably with contemporary cast nickel-base alloys, with the best of the present-day commercial cast alloys. For example, Mar M200 has a 2000°F/15,000 psi stress rupture life of approximately 15 hours, Figure 1. Also, alloys of Series I with high refractory metal content have, in general, average tensile strengths which are comparable to, and in some instances, greater than those of present-day commercial alloys. In addition, the data indicate that low ductility could be a problem at the lower two temperatures, room temperature, and 1400°F.

On the basis of the above results, Mo and Ti were fixed at 2.0 and 1.0 respectively. Ta, W, and Al were scheduled for further study. Ta and W were set at 8.0 and 5.5 respectively, for Series II and III but were studied further in Series IV and V. Al was set at 4.5 for Series II and examined further in Series III.

(b) Series II Alloys (9 compositions)

In Series II, the effects of small additions of the strong carbide formers, Cb, V, and Hf, were studied using the Latin Square design. The levels of the elements used were 0, 1.0, and 2.0 (Square 4 of Table 3).

Statistical evaluations of the 2000°F stress rupture life and of the ductility are shown graphically in Figures 19 and 20 respectively. The results of this analysis show that the elements V and Hf have a statistically significant effect on 2000°F stress rupture life with V having a negative effect, and Hf having a positive one. The elements Cb and Hf appear to have a statistically significant effect on 2000°F stress rupture ductility with Cb having a negative effect, and Hf going through a maximum (positive effect) near 1.0.

Statistical analyses of ultimate tensile strength and percent elongation at temperatures of room temperature, 1400°F, 1875°F, and 2000°F are presented graphically in Figures 21 to 28. These results indicate that the statistically significant effects which each of the elements has on these properties are the following: Cb has a negative effect on room temperature ductility and a positive one on 1400°F and 2000°F tensile strengths. V has a positive effect on 1400°F tensile strength, and a negative one on 2000°F tensile strength above 1.0, and goes through a maximum near 1.0 for 1400°F ductility. Hf has a negative effect on room temperature ductility, and a positive one on 1400°F tensile strength and ductility and on 2000°F tensile strength above 1.0, and goes through a minimum near 1.0 for 1875°F tensile strength, but has a positive effect at the higher level of 2.0.

Examination of the test data, Tables 21 to 25, shows that two of the alloys have an average 2000°F stress rupture life of better than 15 hours, e.g., Alloys d(1.0 Cb, 1.0 Hf) and g (2.0 Cb, 2.0 Hf) with lives of 17.1 and 19.2 hours, respectively. These compare favorably with the best of the present-day commercial cast alloys. In addition, the alloys of Series II exhibit better than average tensile strengths. However, the data indicate that low ductility could be a problem especially at room temperature and 1400°F.

On the basis of the above results, V was eliminated from the next series, while further tests were run on Hf in Series IV. (Series II and III were melted and cast at the same time using the same base compositions so the next series to be run utilizing these results is Series IV.) The element Cb was fixed at 0.5 based on metallurgical judgment.

(c) Series III Alloys (9 compositions)

In Series III, the effects of additions of two not commonly used solid solution strengtheners, Re and Ru, and of variations in Al, the primary gamma-prime former, were examined employing a Latin Square design. The levels of the elements used were as follows: 0, 2.0, and 4.0 for Re; 0, 1.0, 2.0 for Ru; and 4.5, 5.4, and 6.3 for Al (Square 5 of Table 3).

Statistical calculations of the 2000°F stress rupture life and ductility are illustrated graphically in Figures 19 and 20, respectively. The results show that the elements Ru and Al have statistically significant effect on 2000°F stress rupture life with Ru having a negative effect and Al going through a maximum near 5.4. None of the elements appear to have statistical significance on 2000°F stress rupture ductility.

Statistical analyses of ultimate tensile strength and percent elongation at temperatures of room temperature, 1400°F, 1875°F, and 2000°F are shown graphically in Figures 21 to 28. These results indicate that the statistically significant effects which each element has on these properties are as follows: Re has a negative effect on room temperature and 1400°F ductility and a positive one on 1875°F tensile strength. Ru has a negative effect on room temperature ductility and goes through a minimum at about 1.0 for 1875°F tensile strength and ductility. Al has a negative effect on room temperature, 1400°F, 1875°F, and 2000°F tensile strength, and goes through a minimum at about 5.4 for 2000°F ductility.

Examination of the test data, Tables 27 to 31, shows that two of the alloys have an average 2000°F stress rupture life of better than 15 hours; e.g., Alloys d (5.4 Al, 2.0 Re) and g (4.5 Al, 4.0 Re) with lives of 16.6 and 15.5 hours respectively. Like Series II, the tensile strengths are above average but the ductilities are relatively low for Series III alloys.

On the basis of the above results, Ru was eliminated from further consideration in the program, Al was fixed at 5.4 and Re was scheduled for further study in Series IV.

(d) Series IV Alloys (18 compositions)

In Series IV, the effects of four elements not previously studied (Cr, Co, C, and Zr) and four elements requiring further testing (Ta, W, Hf, and Re) were examined using a fractional factorial design in which 8 factors were varied at 2 levels (16 alloys). The design is given in Table 4. The levels of the elements used were as follows: 7.0 and 10.0 for Cr, 5.0 and 10.0 for Co, 0.08 and 0.18 for C, 0.03 and 0.23 for Zr, 8.0 and 10.0 for Ta, 5.5 and 7.5 for W, 1.5 and 2.5 for Hf, and 0 and 1.0 for Re. Two additional alloys were included in this series. One was designed on the basis of past experience in the program and metallurgical judgment, while the other had the design center composition of the fractional factorial experiment, Alloys Y and Z, respectively.

Statistical calculations of the mechanical properties are given in Table 56. It should be noted that both main effects and first order interactions are obtained from the fractional factorial technique. The interactions are confounded; but since there were no statistically significant interaction effects on 2000°F stress rupture life, only the main effects are considered in this discussion. A sample calculation is given in Appendix II.

The main effects calculated from these alloys show that five out of the eight elements varied have statistically significant effects on 2000°F stress rupture life. They are Co, Cr, Ta, W, and Hf. These elements all have a negative effect on life at the element levels tested. There were no statistically significant effects on 2000°F stress rupture ductility.

The statistically significant effects for each of the 8 elements on the tensile properties at the element levels tested are as follows: Co has a negative effect on 1400°F and 2000°F tensile strengths and a positive one on 2000°F ductility. Cr has a negative effect on room temperature and 1400°F tensile strengths and ductility and on 1875°F and 2000°F tensile strengths and a positive one on 2000°F ductility. Ta has a negative effect on 1400°F tensile strength and ductility and on 1875°F and 2000°F ductility. W has a negative effect on room temperature ductility and on 1400°F, 1875°F, and 2000°F tensile strengths. Hf has a negative effect on room temperature and 1400°F tensile strengths. C has a negative effect on 2000°F tensile strength and ductility. Zr has a negative effect on 1875°F and 2000°F tensile strengths. Re has a negative effect on 1400°F tensile strength and on 2000°F tensile strength and ductility.

Examination of the test data, Tables 33 to 37, shows that 7 out of 18 alloys have an average 2000°F stress life of better than 20 hours; e.g., Alloys A, D, J, L, M, N, and Y with lives of 27.3, 26.1, 21.7, 34.0, 23.7, 33.2, and 41.3 hours, respectively. Alloy Y, the alloy designed on the basis of past experience and metallurgical judgment, had the best life of any alloy in the program through Series IV. Again, as in the previous two series, the tensile strengths are above average, and the ductilities are low.

On the basis of the above results, the next series was designed so that it in conjunction with Series IV could be used to determine the constants of a general quadratic equation. This equation would be used for final optimization of four chemical elements. The elements, Cr, Ta, W, and Hf, were chosen from Series IV as being the four (out of eight) elements having the most significant effect on 2000°F stress rupture life. All other elements were fixed at the design-center composition.

(e) Series V Alloys (8 compositions)

In Series V, the effects of Cr, Ta, W, and Hf were examined using a fractional factorial design in which four factors were varied at two levels (8 alloys). The design is given in Table 5. The levels of the elements used were 4.0 and 6.0 for Cr, 6.5 and 7.5 for Ta, 4.0 and 6.0 for W, and 1.00 and 1.75 for Hf.

Statistical calculations of the mechanical properties are given in Table 57. The calculated main effects show that Cr, Ta, and W have statistically significant positive effects. As in Series IV, the interactions were not statistically significant. Ta has a statistically significant negative effect on 2000°F stress rupture ductility.

The statistically significant effects for each of the four elements on the tensile properties are as follows: Cr has a positive effect on 1400°F tensile strength and ductility and on 1875°F and 2000°F tensile strengths. W has a positive effect on 1875°F and 2000°F tensile strengths, but has a negative one on 1875°F and 2000°F ductility. Ta has a positive effect on 1400°F, 1875°F and 2000°F tensile strengths, but has a negative one on room temperature, 1875°F, and 2000°F ductility. Hf has a positive effect on 1875°F and 2000°F tensile strengths, but has a negative one on 2000°F ductility.

The statistical results of Series IV and V for the four elements Cr, Ta, W, and Hf illustrate one of the inherent difficulties in such a two level experiment as noted above. For example, the 2000°F stress rupture life results in Series IV indicated that Cr, Ta, W, and Hf were detrimental to life; but in Series V the results indicated that Cr, Ta, and W were beneficial to life. As can be seen from Tables 2 and 3, the lower levels of Series IV and upper level of Series V for W overlap and those for Cr and Ta are very close. This illustrates the fact that the statistics only indicate the direction one should go, but do not show whether a maximum (or minimum) exists between the two chosen levels or outside them. Therefore, further experimental and/or mathematical results are required to resolve this problem. In this case a mathematical method was used to optimize the effect of these four elements on 2000°F stress rupture life. The quadratic equation obtained from analysis of the 2000°F stress rupture life results of Series IV and V is given below:

$$\begin{aligned}
\mathrm{Log}_{10}\,(2000^{\circ}\mathrm{F}\text{ stress rupture life, hours}) &= 7.088 \qquad (1)\\
&+ 0.604\,(\mathrm{Cr}^{*}) + 1.06(\mathrm{Ta}) + 0.70(\mathrm{W}) + 0.34(\mathrm{Hf})\\
&- 0.30(\mathrm{Cr})^2 - 0.0226(\mathrm{Cr})(\mathrm{Ta}) - 0.0054(\mathrm{Cr})(\mathrm{W}) - 0.006(\mathrm{Cr})(\mathrm{Hf})\\
&- 0.044(\mathrm{Ta})^2 - 0.021(\mathrm{Ta})(\mathrm{W}) - 0.026(\mathrm{Ta})(\mathrm{Hf})\\
&- 0.041(\mathrm{W})^2 - 0.006(\mathrm{W})(\mathrm{Hf})\\
&- 0.041(\mathrm{Hf})^2
\end{aligned}$$

* Concentration of element in parenthesis is in weight percent.

Examination of the test data, Tables 39 to 43, shows that four out of the eight alloys have an average 2000°F stress rupture life of greater than 20 hours; e.g., Alloys A, E, G, and H with lives of 23.3, 21.5, 41.8, and 26.2 respectively. As in previous series, the tensile strengths of the alloys with stress rupture lives of 20 hours or more are above average, while the ductilities are low.

(f) Series VI Alloys (4 compositions)

Series VI was used to optimize further the compositions derived in the program to this point. Alloy A was obtained by maximizing Equation 1. The optimum values calculated for the four elements - Cr, Ta, W, and Hf -were 6.1, 9.0, 5.8, and 0.43, respectively.(Other elements were fixed at design-center levels of Series IV). The remaining three alloys were formulated from a combination of experience gained from the previous alloy series of this program and metallurgical judgment.

Examination of the test data, Tables 45 to 49, shows that all four alloys have an average stress rupture life of greater than 35 hours and two have a life between 60 to 70 hours. Specifically, alloy A, B, C, and D had average lives of 62.9, 39.0, 35.6, and 67.3 hours respectively. Note that the 2000°F stress rupture life predicted for Alloy A by Equation 1 (see Appendix III) is 40.7 hours. The average value actually obtained for Alloy A, 62.9 hours, is much larger than this value. Statistically, this suggests a deficiency in the mathematical model. It appears very likely that terms of higher order than those included in the regression analysis play an important role in determining the strength of nickel-base alloys, even in the restricted range that was studied. As before, the tensile strengths of the alloys are above average while the ductilities are low.

3. Metallographic Analysis

Typical microstructures for some of the alloys in the as-cast condition as derived from the transverse sections of room temperature tensiles are shown in Figures 29 to 34. These photomicrographs illustrate the complexity of the structures developed in the highly alloyed nickel-base systems produced in this program. Figures 29 to 32 show photomicrographs of Alloys IIf, IIIg, IV Y, and VI D. These specimens have 2000°F stress rupture lives of 5.0, 15.5, 41.3, and 67.3 hours respectively. Note that alloys with poor stress rupture lives, IIf and IIIg, have relatively "clean" microstructures. On the other hand, alloys with longer stress rupture lives, IV Y and VI D, contain a considerable amount of precipitate, especially massive gamma-prime. Figure 33 shows electron micrographs (using replica technique) of massive gamma-prime regions of Alloys V G, and VI D. These micrographs show more clearly than the light micrographs that the massive gamma-prime regions are not just large globules of primary gamma-prime but are a much finer dispersion.

Wlodek[18] has shown in IN 100 that the phase which resembles a pearlitic structure within such massive gamma-prime regions is similar to the perovskite carbide reported by Stadelmaier[19]. Similar structures have been reported in U-700[20] and B-1900[21]. These micrographs are typical of much of the massive gamma-prime regions of the better alloys, such as IV Y, V G, VI A, and VI D. Such dispersons of gamma-prime are probably beneficial to the stress rupture life of these alloys. The carbides are largely the massive MC type such as in Figures 34A and B. There also appears to be some grain boundary carbides, possibly $M_{23}C_6$ or M_6C type as seen in Figure 34C. Also note that the secondary gamma-prime of the better alloy, VI D, in Figure 34A is generally larger than that in the poorer alloy, IIIg, in Figures 34B and C.

The above metallographic results are also useful in explaining why there was no significant loss in tensile strengths between room temperature and 1400°F. In fact, in alloys with large quantities of massive gamma-prime, e.g., IV L, IV N, IV Y, and V G, there were substantial increases in tensile strengths from room temperature to 1400°F. Rockwell C hardness tests indicated that this increase was not principally due to strengthening from additional precipitation at the higher temperature, Table 58. The strengthening mechanism in this case appears to be due to a variation in flow characteristics between the gamma-prime phase and the matrix. It has been demonstrated that the flow stress of the gamma-prime phase increases with temperature up to 1400°F[22].

The fracture surface and general appearance of typical stress rupture specimens are shown in Figure 35. The fracture area microstructures of longitudinal sections of stress rupture specimens are shown in Figure 36. These typical specimens indicate that the fractures are intergranular in nature.

4. Workability

Workability in a nickel-base alloy is important if the alloy is to be used as a wrought material. In order to determine the feasibility of adapting the cast compositions developed in this program to wrought procedures, each of the 75 alloys of Task I were subjected to a workability test. The test, which consists of extrusion to bar stock and side pressing, has been discussed previously in the Materials and Procedures section. Typical extrusions are shown in Figure 37. The rating scale used to rate the material is given in Table 11 with typical examples being shown in Figure 38. The results were then analyzed using the same statistical techniques used with the mechanical properties.

In Series I, II, and III, the Latin Square design was used to analyze the property results. In Series I, the effects of varying Mo (1,4.5,8), W (1,5.5,10) and Ta (1,4.58) were studied at three different (Al + Ti) levels (4.5 + 1.0, 6.3 + 1.0, 6.3 + 1.8). Using the base obtained from Series I, Cb (0,1,2), V (0,1,2) and Hf (0,1,2) were examined in Series II; and Re (0,2,4), Ru (0,1,2), and Al (4.5,5.4,6.3) were examined in Series III. The designs are given in Table 3. The statistical evaluation of workability is illustrated graphically in Figures 39 and 40. The statistically significant effects for each element are as follows: in Series I, Mo does not significantly affect workability at any of the three (Al + Ti) levels studies. W has a negative effect at the (6.3 Al + 1.0 Ti) and the (6.3 Al + 1.8) levels. Ta has a negative effect at the (6.3 Al + 1.0 Ti) and (6.3 Al + 1.8 Ti) levels. In Series II, Cb has a negative effect. Neither V nor Hf has a significant effect. In Series III, Re and Al have a negative effect, but Ru does not have a significant effect.

In Series IV and V, the fractional factorial design was used to analyze the property results. In Series IV, the effect of varying Co (5.0 and 10.0), Cr (7.0 and 10.0), Ta (8.0 and 10.0), Hf (1.5 and 2.5), C (0.08 and 0.18), Zr (0.03 and 0.23) and Re (0 and 1.0), and in Series V, the effect of varying Cr (4.0 and 6.0), Ta (6.5 and 7.5), W (4.0 and 6.0), and Hf (1.00 and 1.75) were studied. The designs are given in Tables 4 and 5. The statistical evaluations of workability are given in Tables 59 and 60. The elements of Series IV having a positive effect on workability at the levels tested are as follows (listed in order of decreasing effect): Cr, C, and Zr; and those having a negative effect on workability at the levels tested: W, Re, Hf, and Co. Ta did not affect workability in Series IV. The element of Series V having a positive effect on workability is Hf; and those having a negative effect on workability: Ta, Cr, and W. Again the problem of determining maximum or minimum points noted previously in the mechanical properties section for the two level fractional factorial design is illustrated here.

Examination of the workability results, Table 50 to 55, shows that the alloys with good stress rupture lives are not normally very workable. In general, the elements and levels which increase life and strength decrease workability.

C. Conclusions

Through the use of two statistical techniques, Latin Square and fractional factorial, it has been possible to study the effects of number of elements on the mechanical properties and workability of nickel-base alloys. The elements studied were Ta, W, Mo, Ti, Cb, V, Hf, Re, Ru, Al, Cr, Co, C, and Zr. The level or range of levels of these elements found to produce optimum stress rupture lives with adequate ductility are as follows:

Ta	W	Mo	Ti	Cb	V	Hf	Re
8.0-9.0	5.5-6.0	2.0	1.0	0.5-1.0	0-0.5	0.43-2.0	0-1.0

Ru	Al	Cr	Co	C	Zr
0	5.4	5.4-6.1	5.0-7.5	0.13-0.15	0.03-0.13

Ta, W, Hf, Al, and Cr appear to be the elements most influential in improving stress rupture life. In general, the elements and levels which increase life also increase tensile strength, but they normally decrease ductility and workability.

Based upon the statistical and metallographic results of the 75 experimental alloys of Task I, three cast alloys IV Y, VI A, and VI D, were selected for further study in Task II. The nominal compositions and average 2000°F stress rupture lives (15,000 psi load) of the alloys are given in Table 6. Also the average 2000°F stress rupture lives of the alloys are shown with respect to the target alloy and some present-day alloys in Figure 41. This figure indicates that while the three cast alloys fall slightly short of the proposed target goals of the program, they do represent a substantial improvement of 2000°F stress rupture life over the present-day alloys.

In addition, five wrought alloys, I-5, II b, II d, III g, and IV Y*, were selected from the 75 alloys of Task I for further study in Task II. The basis for the selection of these alloys was a combination of 2000°F stress rupture life and workability rating. The nominal compositions, average 2000°F stress rupture lives, and the workability rating of the five alloys are given in Table 7.

* Note that in the initial selection, Alloys II b, III g, IV L, IV Y, and VI A were chosen. However, due to the inability to obtain sound material from the extrusions of IV L and VI A, I-5 and II d were substituted for them. This will be discussed in more detail in the next section.

VI. EVALUATION OF SELECTED ALLOYS - TASK II

The results obtained in Task II of the program are presented and discussed in this section. The results include melting data, chemical analyses, and mechanical properties.

A. Results

The pertinent melting data for Task II alloys are given in Table 61. The data include melting times and "silverfish" temperatures of virgin heats and casting times and vacuum pressures at 2900°F of remelt heats. The chemical analyses of the cast alloys (three heats for each alloy) are given in Table 62, while those for the wrought alloys (one heat for each alloy) are given in Table 63.

The property results are presented in Tables 64 to 74 for the cast alloys and Table 75 for the wrought alloys. For the cast alloys, the 1400°F, 1875°F, and 1950°F stress rupture results in air are given in Tables 64 to 66, respectively. The 1875°F and 1950°F stress rupture results in an argon atmosphere are presented in Table 67. The stress rupture tables include life in hours, percent elongation, and percent reduction in area. The "as-cast" tensile test data at temperatures of room temperature, 1200°F, 1400°F, 1600°F, 1875°F, and 1950°F are given in Table 68. The tensile test data at temperatures of room temperature, 1200°F, 1400°F, 1600°F, and 1875°F after a 300 hour argon age at 1875°F are presented in Table 69. The tensile test tables include ultimate tensile strength, 0.2% offset yield strength, percent elongation, and percent reduction in area. The results for the unnotched Charpy impact, the thermal fatigue, the oxidation, and the hot corrosion tests are given in Tables 70 to 73, respectively. The results of phase identification by X-ray diffraction of extracted residues are given in Table 74. The stress rupture results for the wrought alloys tested in air at 1875°F are presented in Table 75.

B. Discussion

1. Castability and Machinability

The castability of Task II nickel-base alloys compared favorably with that reported above for Task I alloys. Qualitative observations of the melting and casting operations indicated that there was no difficulty in melting the alloying elements within the limits of the established procedure. Reactive metals were added without violent reactions or metal loss. Fluidity at the pouring temperature was excellent. Recovery of the alloying elements was, in general, very good as is shown by the chemical analyses of the cast and wrought alloys, Tables 62 and 63, respectively.

Further verification of the castability of the cast alloys was the complete absence, with one exception, of porosity detectable by standard radiographic techniques in the cast specimens (i.e., tensile bar, Charpy bar, and thermal fatigue specimens). There was only one specimen rejected in Task II for this reason. No machined specimen was rejected by the dye penetrant test. There was no evidence of cracking or surface checking from machining and/or grinding operations on the tensile and Charpy specimens. Some cracking occurred on grinding the thermal fatigue specimens, but this difficulty was eliminated by first grinding the 1/32 inch flat on the pointed end of the wedge. Therefore, the results obtained in Task II confirm the results of Task I that these alloys possess adequate castability and machinability.

2. Mechanical Property Results

(a) Stress Rupture Results

The stress rupture results, Tables 64 to 67, tend to support the general conclusion drawn from Task I that the three cast alloys represent a distinct improvement in stress rupture life over present-day superalloys. Comparison of the stress rupture lives of the three cast alloys with some present-day superalloys is given in Figures 41 to 44 in terms of an isostress diagram (Figure 41) and Larson-Miller plots (Figures 42 to 44). These results also show that Alloy VI A has the best stress rupture life properties of the three alloys. In addition, comparison of the results in air, Tables 65 and 66, with those in argon, Table 67, indicates that the inert atmosphere was not noticeably beneficial (or detrimental) to the stress rupture properties of these alloys. Therefore, it would appear that oxidation is not a contributing factor to stress rupture failure in these alloys.

The results showed a fair amount of scatter. This scatter could be due to casting segregation or variation in heat chemistry. As a precaution against either possibility, a test bar from each of the three heats was used for each of the test conditions. (That is, the stress rupture results for heat 1 of Alloy IV Y, VI A, or VI D are the first set of results given for the alloy under each test condition (Tables 64 to 67); heat 2, the second set; and heat 3, the third set.) The results indicate that heat chemistry was the primary factor in this scatter, with casting segregation, a secondary one. For example, heats 2 and 3 of Alloy IV Y and VI A and heats 1 and 3 of Alloy VI D generally produced the higher and more consistent results.

Creep curves were developed utilizing platinum or mechanical extensometers and dial gauges. These curves show that the cast alloys go through all three stages of creep before failure. Some examples of these curves are given in Figures 45 to 50. The curves in Figures 45 and 46 are for Alloy IV Y at 1400°F under 85,000 psi and 94,000 psi, respectively.

Similarly, Figures 47 and 48 are for Alloy VI A, and Figures 49 and 50 are for Alloy VI D. These curves were developed using a platinum extensometer. Dial gauge measurements were made for many of the test specimens. The curves developed with the dial gauges showed good correlation with the platinum or mechanical extensometers, once the initial slack in the system was taken up. (The initial Δ l using the dial gauges was greater than that for the extensometers.) However, the dial gauges were only sensitive enough to produce creep curves when the percent elongation was 6-7% or greater. An example of a creep curve developed using a dial gauge and one using a platinum extensometer are given in Figures 51 and 52, respectively.

(b) Tensile Results

The "as-cast" tensile results, Table 68, compare favorably with the results obtained in Task I for these alloys. As before, these results show that an increase in strength occurs over that found at room temperature in the temperature range up to 1400°F - 1500°F. This is illustrated in Figures 53 to 55 where the average ultimate strength and the average percent elongation are plotted as functions of temperature for each alloy. This, as noted previously in the Metallographic Analysis Section of Task I, appears to be due to a variation in flow characteristics between the gamma-prime phase and the matrix[22]. In addition, these alloys have better tensile strengths at all temperatures tested than the present-day commercial alloys[23], some examples of which are given in Table 76. The ultimate tensile strengths of these alloys are, in general, approximately equal to or higher than those reported for high strength commercial cast alloys such as Inco 713C, MAR M-200, and IN 100, and the NASA experimental alloys TAZ-8[24] and TAZ-8A[25]. At all temperatures, the yield strengths at 0.2% offset are generally about 5,000 to 10,000 psi higher than those for the commercial alloys. The ductilities obtained for these alloys are somewhat lower than the average for present-day commercial alloys but appear to be adequate. A slight minimum in tensile elongation was noted for Alloys IV Y and VI A, Figures 53 and 54, respectively. It appears to occur at about 1400°F for IV Y and at about 1600°F for VI A.

The argon aged tensile results (300 hours age at 1875°F), Table 69, indicate that the aging treatment is detrimental to the mechanical properties of the three cast alloys. The average tensile strengths, Table 76, are approximately 10-20% lower; and the ductilities are generally somewhat lower. Metallographic examination of the aged alloys indicates that the decrease in tensile properties was not due to the formation of undesirable phases such as sigma, Laves, or mu. Typical microstructures in the "as-cast" and argon aged conditions are shown in Figures 56 and 57 respectively, for Alloy VI A. (Note that the acicular particles in Figure 57A are M_6C carbides.)

This conclusion was confirmed by residue analysis, Table 74. Metallographic examination, particularly the electron micrographs, Figures 56B and 57B, showed that the probable cause for the decline in tensile properties is the agglomeration of the gamma-prime in the alloys as a result of the aging treatment.

(c) Charpy Impact Results

The results for the unnotched Charpy impact tests are given in Table 70. Comparing the average room temperature results of these alloys (35.0 for IV Y, 28.2 for VI A, and 26.4 for VI D) with the unnotched room temperature results of other nickel-base superalloys, it is found that the results are considerably better than the average value of 9.7 ft.-lbs. for the high-strength cast nickel-base alloy, Nicrotung[24]; but they are somewhat below the values obtained for IN 100 and PDRL 162 (average value range for coarse and fine grain structures - 55 to 75 ft.-lbs.*) These values compare favorably with those obtained for TAZ-8 of 33 ft.-lbs.[24] and TAZ-8A of 24 ft.-lbs.[25].

No values are currently available for comparison of these alloys with other nickel-base alloys at the higher temperatures. However, the results indicate that there is small, if any, change in impact resistance up to about 1400°F.

(d) Thermal Fatigue Results

The three cast alloys were highly resistant to thermal shock, Table 71. At the test temperature of 2100°F, Alloy IV Y had the greatest resistance to thermal shock; and Alloy VI D had the least. The results of some commercial alloys, Inco 713C (virgin), IN 100 (virgin), IN 100 (revert), and WI 52 (air melt) run at 2100°F under analogeous conditions on the TRW rig are also presented in Table 70. It can be seen from these results that Alloys IV Y and VI A are somewhat more thermal fatigue resistant than the commercial alloys. Alloy VI D compares favorably with Inco 713C, the most thermal fatigue resistant of the commercial alloys tested. At the other test temperature of 1875°F, the three alloys ran to 4000 cycles without any cracks being observed. Testing was discontinued at this point. No results for commercial alloys at this test temperature were available for comparison.

* These are International Nickel Co. results supplied through the courtesy of Mr. R. T. Grant.

The above thermal fatigue results appear to reflect the higher strengths of these alloys. This is somewhat surprising considering that the alloys contain a large number of massive gamma-prime particles (e.g., see Figures 31 and 32), which one would normally expect to act as nucleating sites for crack formation. This, in effect, should lower thermal fatigue resistance. Metallographic examination of failed specimens showed that the thermal fatigue cracks did not nucleate or even propagate through or along the boundary of massive gamma-prime particles. Figure 58 shows thermal fatigue cracks in specimens of Alloy VI A and VI D testing at 2100°F.

3. Corrosion Results

(a) Oxidation Results

The oxidation results, Table 72, indicate that two of the three cast alloys, Alloys IV Y and VI A, have good oxidation resistance, with Alloy VI A being the better of the two. Photomicrographs of the metal-oxide interface, depletion zone, and unaffected matrix are given in Figures 59 and 60 for Alloys IV Y and VI A, respectively. The photomicrographs show these zones after 1000 hour exposure times at 1875°F and 2100°F. Alloy VI D, on the other hand, has poor oxidation resistance. Catastrophic oxidation has occurred by the 200 hour exposure time at 2100°F and by the 500 hour time at 1875°F and 1950°F. An example of this type of failure is shown in Figure 61.

It is not possible to make direct comparison of the oxidation data for these alloys with the data for other nickel-base superalloys because of the variations in experimental techniques used by different investigators. These results in conjunction with observations from the 1875°F, 1950°F, and 2000°F stress-rupture specimens indicate that Alloys IV Y and VI A are more oxidation resistant than MAR M-200, IN 100, TAZ8, and Rene' 41 and are comparable to vacuum melted TAZ8A[25].

Several anomalies appeared in the oxidation tests. For Alloy VI A at 2100°F and 500 hours, the specimen surface in contact with the crucible appeared to have melted. This resulted in the weight gained at 500 hours being greater than that at 1000 hours. This anomaly was probably due to contamination of the material either by the crucible or some impurity which entered the crucible during the test, thereby producing a low melting phase. The result should be disregarded. Some surface melting was also noted on Alloy IV Y specimens after exposure times of 500 and 1000 hours at 2100°F. This is probably due to the formation of relatively low melting oxides. Finally, some of the results, particularly for Alloy IV Y, showed a lower weight gained at the higher temperatures than at the lower temperatures for corresponding exposure times. The probable cause for this anomaly is evaporation of the alloying constituents.

(b) Hot Corrosion Results

The hot corrosion (or sulfidation) results, Table 73, show that Alloy VI A has the best hot corrosion behavior of the three cast alloys under the test conditions using the crucible test and that Alloy VI D has the worst. The results were similar to those found for the oxidation behavior of these alloys.

The results of a number of commercial superalloys run under analogous conditions are compared with the cast alloys in Figure 62.* It is evident from this figure that Alloys IV Y and VI A have somewhat better hot corrosion resistance than other high strength nickel-base alloys, such as MAR M-200, B 1900, TRW 1900 and Alloy VI D. The hot corrosion behavior of Alloy IV Y is comparable to that of Inco 713C; while the behavior of Alloy VI A, is comparable to that of IN 100 and PDRL 163.

4. Phase Analysis Results

Phase analysis studies were carried out on five alloy conditions in order to identify the secondary phases present. The alloy conditions include the three cast alloys in the "as-cast" condition and Alloy VI A after a 1875°F, 15,000 psi stress-rupture test and after a 300 hour argon age at 1875°F. The results, Table 74, indicate that the secondary phases include only carbides and borides. (Note that gamma-prime is considered to be a primary phase in these alloys.) The major secondary phase present for the five alloy conditions is the MC type carbide, as indicated by the strong or very strong X-ray diffraction lines. A minor secondary phase present for the five alloy conditions is the M_3B_2 boride, as indicated by the weak or very weak X-ray diffraction lines. A small amount of M_6C type carbide was found in Alloy VI A after being aged at 1875°F for 300 hours in argon and in the "as-cast" condition for Alloy VI D. While it was not possible to positively identify its presence, there were some weak or very weak lines which indicated that M_6C may also be present in the "as-cast" condition for Alloys IV Y and VI A. Also, a small amount of $M_{23}C_6$ type carbide was identified in the "as-cast" condition of Alloy VI D.

* It should be noted that the relative rankings of the commercial nickel-base alloys using the crucible test, Figure 62, agree favorably with those obtained by the Allison Division of General Motors on a burner rig. This information was provided by Mr. Ken Ryan of the Allison Division of General Motors.

In addition to the above alloy conditions, phase analysis was utilized in conjunction with a 1500 hour age at 1600°F to evaluate the microstructural stability of the three alloys. The object was to determine if these alloys will form phases such as sigma, Laves, or mu, which are detrimental to mechanical properties. It appears from the results of this analysis, Table 74, that these alloys do not suffer from a microstructural instability problem. The results show that the only secondary phases present are carbides (primarily MC type) and M_3B_2 borides. A typical microstructure after the 1600°F/1500 hr. age is shown in Figure 63 for Alloy VI A.

Electron-vacancy calculations* were also performed for the three alloys. The electron vacancy numbers obtained from these calculations indicated that the alloys were very "safe", i.e., sigma will not form. For example, using the method espoused by G.E. called PHACOMP with a gamma-prime composition of Ni_3 (Al, Ti, Cb, Ta, 0.03 original atomic percent Cr)[26], electron-vacancy numbers of 2.10, 2.11, and 2.04 were obtained for Alloys IV Y, VI A, and VI D, respectively. The critical value for this method is 2.49, i.e., the value below which sigma will not form.

5. Wrought Alloys

Initially, the five alloys selected for study in Task II included Alloys IIb, IIIg, IV L, IV Y, and VI A. These alloys were to be double extruded from nominal 3 inch cast ingots to nominal 1/2 inch bar. However, two of the alloys, IV L and VI A, were found to be unacceptable for further working after the primary extrusion due to the lack of sound material. As a result, two different alloys, I-5 and IId, were substituted for them. These five alloys were successfully double extruded to nominal 1/2 inch bar.

A heat treatment was then developed for each alloy to attempt to optimize the 1875°F stress rupture properties. The development of the heat treatment consisted of first determining a solutioning temperature such that the gamma-prime is dissolved without excessive grain growth or preferential liquation (incipient melting). A study of the solution temperatures over the range of 2175°F to 2325°F (for four hours) was made for the five alloys. As a result of this study, a solution temperature of 2250°F was selected for four of the five alloys, I-5, IIb, IId, and IIIg. The solution temperature selected for Alloy IV Y was 2290°F. It should be noted that it was not possible to completely solution the gamma-prime before incipient melting occurred in Alloy IV Y. So the above solutioning temperature represents an attempt to compromise between gamma-prime solutioning and incipient melting. Then the heat treatment was completed with a standard heat treating cycle of 4 hours at 2000°F, 24 hours at 1550°F, and 16 hours at 1400°F.

* All electron-vacancy methods which have been used in an Air Force Program at TRW entitled "Research on Microstructural Instability on Nickel-Base Superalloys," Contract AF 33(615)-5126 have been employed.

The stress rupture results, Table 75, show that four of the five alloys (I-5, IIb, IId, and IIIg) look promising as wrought alloys. For example, at 1875°F and 10,000 psi, U-700, the best high temperature commercially available wrought superalloy, has a stress rupture life of approximately 100 hours(23). Under these test conditions, Alloy I-5, IIb, IId, and IIIg have average lives ranging from approximately 245 hours for Alloy IIb to approximately 320 hours for Alloy IIIg. However, comparing the stress rupture results of these alloys with those obtained for the cast alloys under the same test conditions, Table 65, it can be seen that the wrought alloys have considerably lower stress rupture lives than the cast alloys (approximately an order of magnitude lower). Note also that the ductilities for these alloys are low for wrought alloys.

Alloy IV Y, however, shows very little promise as a wrought alloy. Its problem is one suffered by most highly alloyed materials, i.e. the gamma-prime cannot be sufficiently solutioned before incipient melting occurs. Hence, the worked structure cannot be properly heat treated resulting in poor mechanical properties.

6. Metallographic Analysis

(a) Cast Alloys

Typical microstructures for the three alloys in the "as-cast" condition as derived from the transverse sections of room temperature tensiles are shown in Figures 31 and 32 for Alloys IV Y and VI D, respectively, and in Figure 56 for Alloy VI A. The effect of such microstructures on the mechanical properties was discussed previously in the metallographic analysis section of Task I.

The fracture surface and general appearance of typical stress rupture specimens are shown in Figure 64. These specimens illustrate the relatively brittle nature of the fractures found in the three alloys for all stress rupture test conditions.

The fracture area microstructures of typical stress rupture specimens as derived from longitudinal sections are shown in Figures 65 to 68. These figures indicate that the fractures are intergranular in nature at all stress rupture test conditions. Also, Figures 66 to 68 show that gamma-prime is forming columnar grains at the higher test temperatures of 1875°F and 1950°F. The agglomeration of the gamma-prime in this manner is probably detrimental to the high temperature stress rupture properties of these alloys. An analogous problem was noted previously for the decline in tensile properties after a 300 hour argon age at 1875°F.

(b) Wrought Alloys

A typical microstructure of a nickel-base alloy (Alloy IIIg) in the double extruded condition is shown in Figure 69A. This figure shows that the breakdown from the cast structure, while not complete is considerable. There are remnants of the cast dendritic structure present as indicated by the elongated granular structure and general inhomogeneity. There are also some primary gamma-prime pools present.

A typical microstructure of heat treated nickel-base alloys (Alloy IIIg) is given in Figure 69B. This figure shows that some carbide are present in "chains". Some of the carbides probably did not go into solution, and they retained approximately the same morphology that they had before the heat treatment. In addition, some fine carbide formed during heat treatment, often in the carbide "chains", indicating the presence of segregation even after solutioning. In these highly alloyed materials, the gamma-prime assumes a partially overaged aged appearance (the speckled effect in the grains.) In some alloys there is also some primary gamma-prime present in the form of large "pools" (not shown in figure).

It was noted previously that the solutioning temperature used for Alloy IV Y represented an attempt to compromise between gamma-prime solutioning and incipient melting. Microstructural examination of the heat-treated material, Figure 70, indicated that some incipient melting did occur during solutioning. However, since no appreciable increase in stress rupture life could be expected from a lower solutioning temperature (because of incomplete solutioning of gamma-prime), the material was tested in this heat-treated condition.

The fracture area microstructures of typical stress rupture specimens as derived from longitudinal sections are shown in Figure 71. This figure indicates that the fractures in the wrought alloys are analogous to those in the cast alloys, i.e., intergranular in nature for all test conditions.

C. Conclusions

Task II entailed a more complete property evaluation of the three most promising cast alloys and the five most promising wrought alloys obtained in Task I. The following conclusions were derived from this evaluation.

For the cast alloys, Alloys IV Y, VI A, and VI D, the stress rupture results of Task II substantiated the general conclusions drawn from Task I that the three alloys represent a substantial improvement in stress rupture life over present-day alloys, Figures 41 to 44. Alloy VI A has the best combination of high temperature properties of the three alloys. Its stress rupture life is the highest and represents approximately a 50°F increase in temperature over the lives of present-day high strength nickel-base alloys. Its tensile strength , impact properties, and density (0.306 lbs/cu in.) are comparable with such alloys. Its ductility, while below average, appears to be adequate; and it does not appear to suffer a severe reduction in tensile ductility in the 1400°F to 1600°F temperature range, Figure 54. In addition, Alloy VI A has good thermal fatigue and corrosion resistance for a high-strength alloy and does not appear to suffer from microstructural instability, e.g., sigma, Laves, or mu formation. A major disadvantage of this alloy appears to be that of cost. The rhenium in the alloy adds about $3.00 per pound to the cost of the virgin material. Further work is suggested to determine whether the amount of rhenium can be reduced or even eliminated without unduly affecting the properties of the alloy.

For the wrought alloys, four of the five alloys, Alloys I-5, IIb, IId, and IIIg, show promise as wrought alloys. The 1875°F stress rupture results for these alloys show a substantial improvement in life over wrought U-700. However, further property evaluations are required to determine the full potential of the four wrought alloys.

VII. REFERENCES

1. A. Roy, F.A. Hagen, and J.M. Cowin, "Iron-Base Superalloys for Turbine Engines", Journal of Metals, 17, 934, September 1965.

2. J.C. Freche, R.L. Ashbrook, and S.J. Klima, "Cobalt-Base Alloys for Space - Power Systems", Journal of Metals, 15, 928 (1963).

3. J.C. Freche, and W.J. Waters, "High Temperature Service Offered by New Nickel-Base Alloys", Foundry 92, 44, July 1964.

4. J.B. Newkirk, "General Theory Mechanism and Kinetics", Precipitation from Solid Solution, ASM, Cleveland, Ohio, 1959.

5. R.F. Decker and R.R. DeWitt, "Trends in High-Temperature Alloys", Journal of Metals, 17, 139, 1965.

6. C.H. Lund, "Physical Metallurgy of Nickel-Base Superalloys", Defense Metals Information Center, Battelle Memorial Institute, DMIC Report No. 153, May 5, 1961.

7. R.G. Davis and N.S. Stoloff, "On the Yield Stress of Aged Ni-Al Alloys", Trans, Met, Soc. AIME, 233, 714, 1965.

8. M. Hansen and K. Anderko, Constitution of Binary Alloys, 2nd ed., McGraw-Hill, New York, 1958.

9. L.R. Woodyatt, C.T. Sims, and H.J. Beattie, Jr., "Prediction of Sigma-Type Phase Occurrence from Compositions in Austenitic Superalloys", Trans. Met. Soc. AIME, 236, 519 1966.

10. R.F. Decker and J.W. Freeman, "The Mechanism of Beneficial Effects of Boron and Zirconium on Creep Properties of a Complex Heat Resistant Alloy", Trans. Met. Soc. AIME, 218, 277, 1960.

11. C.P. Sullivan and M.J. Donachie, Jr., "Some Effects of Microstructure on the Mechanical Properties of Nickel-Base Superalloys", Metals Engineering Quarterly, 7, 36, February 1967.

12. B.S. Natapov, V.E. Ol'shanetskii and E.P. Ponomarenko, "Effect of Alloyed Elements on the Form of Secondary Precipitates in Refractory Nickel Alloys", Metal Science and Heat Treatment, pp. 11-14, January - February 1965.

13. W.J. Dixon and F.J. Massey, Introduction to Statistical Analysis, McGraw-Hill, New York, 1951.

14. J.F. Barker, "1400°F Ultra-High Strength Alloy Development Program", Interim Progress Report #1, Contract No. AF 33(615)-1597, October 15, 1964.

15. R.L. Plackett and J.P. Burman, "The Design of Optimum Multifactorial Experiments", Boimetrika, 33, 305, 1946.

16. M.G. Natrella, Experimental Statistics, National Bureau of Standards Handbook 91, United States Department of Commerce, August 1963.

17. O.L. Davies (Ed.), The Design and Analysis of Industrial Experiments, Hafner Publishing Co., New York, New York, 1954.

18. S.T. Wlodek, "The Structure of IN-100", ASM Trans. Quarterly, 57, 110, 1964.

19. H.M. Stadelmaier, "Ternary Compounds of Transition Metal, B-Metal and Metalloid", Z for Metallkunde, 52, 758, 1961.

20. J.F. Radavich and W.H. Couts, "Effect of Temperature Exposure on the Microstructure of 4.5%-3.5%Ti Nickel-Base Alloy", ASM Trans. Quarterly, 54, 591, 1961.

21. W.P. Danesi, M.J. Donachie, and J.F. Radavich, "Phase Reactions in B-1900 Nickel-Base Alloy from 1600 to 1800°F, ASM Trans. Quarterly, 59, 505, 1966.

22. R.G. Davies and N.S. Stoloff, "On the Yield Stress of Aged Ni-Al Alloys", Trans. Met. Soc. AIME, 233, 714, 1965.

23. The International Nickel Company, Inc., Nickel Base Alloys, revised edition (1964).

24. John C. Freche, and William J. Waters, "Continued Investigation of an Advanced-Temperature, Tantalum-Modified, Nickel-Base Alloy", NASA TN D-1531(1963).

25. William J. Waters and John C. Freche, "Investigation of Columbium-Modified NASA TAZ-8 Superalloy", NASA TN D-3597 (1966).

26. Chester T. Sims, "A Contemporary View of Nickel-Base Super-Alloys", Journal of Metals, 18, 1119, (October 1966).

TABLE 1

Nominal Compositions of Contemporary Nickel-Base Superalloys

Alloy	Cast or Wrought	C	Cr	Mo	Ti	Al	Co	W	Zr	B	Ta	Cb	V	Fe
Nimonic 75	W	0.04	20.0	-	-	-	-	-	-	-	-	-	-	-
Inconel X	W	0.04	15.0	-	2.5	0.8	-	-	-	-	-	0.9	-	6.8
Waspaloy	W	0.08	19.5	4.3	3.0	1.3	13.5	-	0.06	.006	-	-	-	-
René 41	W	0.09	19.0	10.0	3.1	1.5	11.0	-	-	.005	-	-	-	-
Udimet 700	W	0.08	15.0	5.0	3.5	4.3	18.5	-	.005	.025	-	-	-	-
Inco 713C	C	0.12	12.5	4.2	0.8	6.1	-	-	0.10	.012	-	2.0	-	-
IN 100	C	0.18	10.0	3.0	4.7	5.5	15.0	-	0.06	.014	-	-	1.0	-
Mar M-200	C	0.15	9.0	-	2.0	5.0	10.0	12.5	0.05	.015	-	1.0	-	-
TAZ 8	C	0.125	6.0	4.0	-	6.0	-	4.0	1.0	.004	8.0	-	2.5	-
TAZ8A	C	0.125	6.0	4.0	-	6.0	-	4.0	1.0	.004	8.0	2.5	-	-

TABLE 2

Stress Rupture Properties of Contemporary Alloys

A. 1800°F Tests

Wrought Nickel-Base Alloys	Hours of Life for Given Stress, 10^3 psi: 10	100	1000	10,000
Rene' 41	17	9		
Udimet 700	26	16	7	

Cast Nickel-Base Alloys	10	100	1000	10,000
Inco 713C	29	21	13	8
IN 100		25	15	
MAR M200	37	28	20	15
TAZ 8	35	23	15(1815°F)	

B. 1900°F Tests

Wrought Nickel-Base Alloys	10	100	1000	10,000
Udimet 700	17	8	3	

Cast Nickel-Base Alloys	10	100	1000	10,000
Inco 713 C		12		
IN 100		16	9	
MAR M200	23	18	11	8
TAZ 8		15(1915°F)		
TAZ 8A		15(1895°F)		

TABLE 3

Latin Square Designs Used in Series I, II, and III

Square 1 (Series I)

4.5* Al + 1.0 Ti
Vary Mo: 1, 4.5, and 8
Vary W: 1, 5.5, and 10
Vary Ta: 1, 4.5, and 8

Square 2 (Series I)

6.3 Al + 1.0 Ti
Same Mo, W, and Ta levels as for Square 1

Square 3 (Series I)

6.3 Al + 1.8 Ti
Same Mo, W, and Ta levels as for Square 1

Square 4 (Series II)

Vary Cb: 0, 1, and 2
Vary V: 0, 1, and 2
Vary Hf: 0, 1, and 2

Square 5 (Series III)

Vary Re: 0, 2, and 4
Vary Ru: 0, 1, and 2
Vary Al: 4.5, 5.4, and 6.3

* Compositions are in wt. %

TABLE 4

Fractional Factorial Design, Elements, Design Symbols, and Levels for Series IV

Plan 8 - Sixteen Experiments: Treatment Combinations, Unmodified (from Ref.16)	Modified	Alloy
1 *	gh	L
aegh **	ae	A
befg	befh	C
abfh	abfg	O
cefh	cefg	M
acfg	acfh	D
bcgh	bc	G
abce	abcegh	P
dfgh	df	N
adef	adefgh	J
bdeh	bdeg	F
abdg	abdh	K
cdeg	cdeh	I
acdh	acdg	H
bcdf	bcdfgh	E
abcdefgh	abcdef	B

Element	Design Symbol for Element	Levels (Wt. %) (1)	(0)
Co	a	10.0	5.0
Cr	b	10.0	7.0
Ta	c	10.0	8.0
Hf	d	2.5	1.5
W	e	7.5	5.5
Re	f	1.0	0
C	g	0.18	0.08
Zr	h	0.23	0.03

* Signifies that all elements are at the (0) level.

** Signifies that elements with the design symbols listed are at the (1) level, all other elements are at the (0) level.

TABLE 5

Fractional Factorial Design, Elements, Design Symbols, and Levels for Series V

Plan 2 - Eight Experiments (from Ref. 16)			Design Symbol for Element	Levels (Wt. %)	
Treatment Combinations	Alloy	Element		(1)	(0)
(1) *	D	Cr	a	6.0	4.0
ad **	H	W	b	4.0	6.0
bd	C	Ta	c	7.5	6.5
ab	B	Hf	d	1.0	1.75
cd	A				
ac	G				
bc	F				
abcd	E				

* Signifies that all elements are at the (0) level.

** Signifies that elements with the design symbols listed are at the (1) level, all other elements are at the (0) level.

TABLE 6

Three Cast Alloys for Task II

	Nominal Composition														
Alloy	C	Cr	Mo	Ti	Al	Co	W	Re	Hf	Zr	B	Ta	Cb	V	Ni
IV Y	0.15	6.0	2.0	1.0	5.4	5.0	5.5	1.0	2.0	0.03	0.02	8.0	1.0	0	Bal.
VI A	0.13	6.1	2.0	1.0	5.4	7.5	5.8	0.5	0.43	0.13	0.02	9.0	0.5	0	Bal.
VI D	0.15	5.4	2.0	1.0	5.4	5.0	6.0	0	1.75	0.08	0.02	8.5	0.5	0.5	Bal.

Alloy	Average 2000°F, 15,000 psi Stress Rupture Life (hours)
IV Y	41.3
VI A	62.9
VI D	67.3

TABLE 7

Five Wrought Alloys for Task II

Alloy	Nominal Composition														
	C	Cr	Mo	Ti	Al	Co	W	Re	Hf	Zr	B	Ta	Cb	V	Ni
I-5	0.13	10.0	1.0	1.0	6.3	10.0	5.5	-	-	0.03	0.02	4.5	-	-	Bal.
II-b	0.13	10.0	2.0	1.0	4.5	10.0	5.5	0	1.0	0.03	0.02	8.0	0	1.0	Bal.
II-d	0.13	10.0	2.0	1.0	4.5	10.0	5.5	1.0	-	0.03	0.02	8.0	1.0	-	Bal.
III-g	0.13	10.0	2.0	1.0	4.5	10.0	5.5	4.0	0	0.03	0.02	8.0	0	0	Bal.
IV-Y	0.15	6.0	2.0	1.0	5.4	5.0	5.5	1.0	2.0	0.03	0.02	8.0	1.0	0	Bal.

Alloy	Average 2000°F 15,000 psi Stress Rupture Life (Hours)	Workability Rating *
I-5	11.9	3.5
II-b	14.3	3.0
II-d	17.1	3.5
III-g	15.5	3.0
IV-Y	41.3	4.0

* See Table 11 for definition of Rating Scale.

TABLE 8

Purity Level for Raw Materials Used in Melting of Nickel Alloys

Addition	Minimum Purity, Weight Percent
Aluminum	99.8*
Chromium	99.5
Cobalt	99.5
Hafnium	98.5**
Nickel Boride	99.0
Titanium	99.8
Vanadium	99.7
All Other Additions	99.9

* For the first 27 heats (Series I) of Task I, the minimum purity of aluminum was 99.7%.

** Major impurity was zirconium, approximately 1.5% - 2.0%.

TABLE 9

Melting Procedure for Virgin Heats

Charging

1. Charge in Crucible (Magnorite) Ni, Co, Cr, W, Mo, Ta, Cb, Re, V, Ru*
2. Charge in first hopper
   Carbon (AUC grade).
3. Charge in second and third hoppers
   Aluminum.
4. Charge in fourth hopper
   Ti, Zr, NiB, Hf.

Melting

1. Pump down to 10μ pressure (maximum).
2. Turn power on -melt down.
3. Take to 3000°F -hold until bath is quiet.
4. Cool to form skin.
5. Add carbon and melt in slowly.
6. After boil ceases (constant power input), and pressure comes down (below 10μ) add aluminum.
7. After aluminum is melted in (2-5 minutes) add Ti, Zr, NiB, Hf.
8. Stir under low power for 5 minutes.
9. Cool until crystals appear - record temperature.
10. Adjust temperature to 2550° - 2575°F or at least 50°F above freezing temperature, whichever is higher and pour.
11. Cure 10 minutes under vacuum.

*For alloys 3b and 3c ruthenium was added from a charging hopper in step 6 after the carbon boil. Rhenium and ruthenium powder was wrapped in Al and Ni foil prior to charging.

TABLE 10

Melting Procedure for Test Bar Clusters

1. Preheat molds for 4 hours in air at 1700°F.
2. Charge "master" ingot into crucible liner (Taycor).
3. Position the preheated mold on the furnace turn-table.
4. Pump down to 10μ pressure (maximum).*
5. Melt and heat to 2900°F.
6. Cool to 2825°F and pour into cluster molds.
7. Cure 5-15 minutes under vacuum.
8. Remove from investment and mold after 2 hours.

* For Series I alloys power was turned on at the start of pump down.

TABLE 11

Rating Scale for Workability Studies

Rating	Definition*
1	Completely workable, e.g., Waspaloy
2	Workable by best forging technique, e.g., U 700
3	Workable by special techniques, e.g., IN 100
4	Very limited workability
5	Unworkable by forging
6	No sound material obtained from extrusion

* See Figure 38 for typical examples.

TABLE 12

Melting Procedure for Extrusion Ingots

1. Charge "master" ingot into crucible liner (Taycor).

2. Position mold on the furnace turn-table.

3. Pump down to 10 μ pressure (maximum).

4. Melt and heat to 2900°F.

5. Cool to 2600°F and pour into "Flower Pot" type mold.

6. Cure 5 minutes under vacuum.

7. Remove ingot from mold and cool in vermiculite for 2 hours.

TABLE 13

Pertinent Melting Data for Task I

Alloy	Virgin Heats: Total Time, hrs.	Virgin Heats: "Silverfish" Temperature, °F	Re-Melt Heats: Total Time, hrs.	Re-Melt Heats: Pressure at 2900°F, μ
Series I				
1	1:16	2550	0:16	7
2	1:12	2520	0:23	6
3	1:05	2475	0:12	6
4	1:00	2500	0:12	6
5	0:54	2480	0:12	8
6	1:06	2450	0:11	6
7	1:26	2460	0:10	6
8	1:15	2450	0:14	8
9	0:57	2550	0:13	6
10	1:07	2470	0:12	6
11	0:53	2500	0:10	7
12	0:58	2470	0:17	4
13	1:06	2480	0:12	7
14	0:52	2490	0:12	7
15	1:10	2400	0:14	5
16	1:08	2480	0:14	6
17	1:07	2440	0:16	6
18	N.A.*	2420	0:11	7
19	1:06	2440	0:10	7
20	0:59	2420	0:14	7
21	1:00	2500	0:12	7
22	0:46	2450	0:16	N.A.
23	1:12	2460	0:14	7
24	1:10	2440	0:12	6
25	0:54	2430	0:11	6
26	0:55	2480	0:12	6
27	1:15	2490	0:12	7
Series II				
a	1:17	2450	0:26	8
b	1:12	2490	0:20	5
c	1:01	2440	0:21	5
d	1:13	2460	0:22	4
e	1:00	2470	0:21	4
f	1:12	2430	0.23	4
g	1:26	2425	0:26	3
h	1:05	2430	0:25	4
i	1:18	2425	0:18	4

* Not available

TABLE 13 (Continued)

Pertinent Melting Data for Task I

Alloy	Virgin Heats: Total Time, hrs.	Virgin Heats: "Silverfish" Temperature, °F	Re-Melt Heats: Total Time, hrs.	Re-Melt Heats: Pressure at 2900°F, μ
Series III				
a	1:18	2430	0:22	5
b	1:12	2460	0:22	4
c	1:06	2475	0:21	4
d	1:13	2500	0:21	4
e	0:57	2480	0:22	4
f	0:52	2480	0:21	4
g	1:11	2450	0:21	6
h	0:49	2520	0:20	4
i	0:53	2510	0:26	4
Series IV				
A	1:14	2440	0:17	1
B	0:55	2370	0:15	1
C	0:55	2375	0:13	1
D	1:04	2425	0:18	1
E	1:10	2460	0:12	2
F	1:10	2490	0:15	2
G	1:04	2420	0:13	1
H	1:08	2420	0:14	1
I	1:10	2390	0:13	2
J	1:04	2440	0:16	1
K	1:07	2390	0:12	1
L	1:13	2440	0:15	1
M	1:21	2400	0:14	2
N	1:07	2410	0:15	1
O	1:12	2410	0:14	1
P	1:00	2430	0:13	1
Y	1:16	2425	0:14	1
Z	0:53	2390	0:16	2
Series V				
A	1:09	2475	0:21	2
B	0:53	2455	0:21	2
C	0:59	2480	0:21	2
D	0:49	2450	0:21	2
E	0:54	2475	0:19	2
F	1:15	2475	0:20	2
G	1:20	2460	0:20	2
H	0:56	2460	0:19	2
Series VI				
A	1:08	2490	0:21	5
B	0:57	2510	0:20	4
C	0:57	2505	0:17	4
D	1:04	2510	N.A. *	N.A.

* Not available.

TABLE 14

Aim and Actual Chemical Compositions for Series I Alloys*

Alloy	Latin Sq. No.	Components Varied at Three Levels: Mo	W	Ta	Al +	Ti	Fixed Components: Co	Cr	Zr	B	C	Ni
I-1	1	1.0 1.04	1.0 1.14	1.0 1.05	4.5 3.99	1.0 1.13	10.0 10.20	10.0 9.79	0.03 0.07	0.02 0.017	0.13 0.09	Bal. -
I-2	2	1.0 1.03	1.0 1.02	1.0 1.00	6.3 6.45	1.0 1.09	10.0 10.15	10.0 9.70	0.03 0.06	0.02 0.019	0.13 0.10	Bal. -
I-3	3	1.0 1.05	1.0 1.05	1.0 1.07	6.3 5.93	1.8 1.89	10.0 10.20	10.0 9.86	0.03 0.05	0.02 0.024	0.13 0.11	Bal. -
I-4	1	1.0 1.07	5.5 5.52	4.5 4.69	4.5 4.72	1.0 1.03	10.0 10.10	10.0 9.64	0.03 0.04	0.02 0.017	0.13 0.12	Bal. -
I-5	2	1.0 1.07	5.5 5.18	4.5 4.54	6.3 6.41	1.0 1.02	10.0 10.16	10.0 9.76	0.03 0.05	0.02 0.030	0.13 0.13	Bal. -
I-6	3	1.0 1.05	5.5 5.67	4.5 4.40	6.3 6.37	1.8 1.86	10.0 10.18	10.0 9.72	0.03 0.05	0.02 0.018	0.13 0.11	Bal. -
I-7	1	1.0 1.03	10.0 10.06	8.0 7.95	4.5 4.50	1.0 1.02	10.0 10.10	10.0 9.80	0.03 0.05	0.02 0.015	0.13 0.13	Bal. -
I-8	2	1.0 1.05	10.0 9.95	8.0 7.90	6.3 6.29	1.0 1.03	10.0 10.15	10.0 9.71	0.03 0.03	0.02 0.015	0.13 0.14	Bal. -
I-9	3	1.0 1.11	10.8 9.84	8.0 7.95	6.3 6.34	1.8 1.81	10.0 10.10	10.0 9.68	0.03 0.04	0.02 0.019	0.13 0.13	Bal. -
I-10	1	4.5 4.55	1.0 1.08	4.5 4.27	4.5 4.52	1.0 1.05	10.0 10.00	10.0 9.76	0.03 0.02	0.02 0.017	0.13 0.13	Bal. -

* Aim is shown first. All data is in weight percent.
Trace Elements: Fe <0.30, Mn <0.05, Si <0.15, and S <0.01.

TABLE 14 (Continued)

Aim and Actual Chemical Composition for Series I Alloys*

Alloy	Latin Sq.No.	Components Varied at Three Levels					Fixed Components					
		Mo	W	Ta	Al +	Ti	Co	Cr	Zr	B	C	Ni
I-11	2	4.5 4.51	1.0 1.01	4.5 4.63	6.3 6.27	1.0 1.00	10.0 10.00	10.0 9.79	0.03 0.04	0.02 0.021	0.13 0.13	Bal. -
I-12	3	4.5 4.28	1.0 0.66	4.5 4.29	6.3 4.60	1.8 1.89	10.0 10.14	10.0 9.94	0.03 0.06	0.02 0.021	0.13 0.13	Bal. -
I-13	1	4.5 4.61	5.5 5.63	8.0 7.91	4.5 4.79	1.0 1.03	10.0 10.20	10.0 9.68	0.03 0.02	0.02 0.017	0.13 0.12	Bal. -
I-14	2	4.5 4.46	5.5 5.21	8.0 7.74	6.3 6.37	1.0 1.03	10.0 10.08	10.0 9.65	0.03 0.03	0.02 0.020	0.13 0.12	Bal. -
I-15	3	4.5 4.50	5.5 5.58	8.0 7.82	6.3 6.32	1.8 1.91	10.0 10.00	10.0 9.70	0.03 0.05	0.02 0.017	0.13 0.12	Bal. -
I-16	1	4.5 4.55	10.0 9.68	1.0 1.05	4.5 4.50	1.0 1.03	10.0 10.13	10.0 9.78	0.03 0.04	0.02 0.018	0.13 0.13	Bal. -
I-17	2	4.5 4.61	10.0 10.45	1.0 1.21	6.3 6.45	1.0 1.09	10.0 10.11	10.0 9.61	0.03 0.05	0.02 0.017	0.13 0.12	Bal. -
I-18	3	4.5 4.58	10.0 9.95	1.0 1.06	6.3 6.05	1.8 1.86	10.0 10.02	10.0 9.75	0.03 0.05	0.02 0.017	0.13 0.13	Bal. -
I-19	1	8.0 7.92	1.0 1.09	8.0 7.92	4.5 4.54	1.0 1.05	10.0 10.05	10.0 9.71	0.03 0.04	0.02 0.015	0.13 0.13	Bal. -
I-20	2	8.0 7.89	1.0 1.21	8.0 7.49	6.3 6.24	1.0 1.05	10.0 10.13	10.0 9.79	0.03 0.05	0.02 0.019	0.13 0.12	Bal. -

* Aim is shown first. All data is in weight percent.
Trace Elements: Fe <0.30, Mn <0.05, Si <0.15, and S <0.01.

TABLE 14 (Continued)

Aim and Actual Chemical Composition for Series I Alloys*

Alloy	Latin Sq. No.	Components Varied at Three Levels					Fixed Components					
		Mo	W	Ta	Al	+ Ti	Co	Cr	Zr	B	C	Ni
I-21	3	8.0 7.98	1.0 0.76	8.0 7.73	6.3 6.27	1.8 1.92	10.0 10.18	10.0 9.69	0.03 0.05	0.02 0.021	0.13 0.13	Bal. -
I-22	1	8.0 7.96	5.5 5.54	1.0 1.06	4.5 4.52	1.0 1.05	10.0 10.06	10.0 9.83	0.03 0.04	0.02 0.020	0.13 0.13	Bal. -
I-23	2	8.0 7.98	5.5 5.35	1.0 1.14	6.3 6.20	1.0 1.06	10.0 10.12	10.0 9.60	0.03 0.03	0.02 0.021	0.13 0.12	Bal. -
I-24	3	8.0 7.98	5.5 5.17	1.0 1.06	6.3 6.47	1.8 1.92	10.0 10.06	10.0 9.53	0.03 0.06	0.02 0.021	0.13 0.10	Bal. -
I-25	1	8.0 7.92	10.0 10.20	4.5 4.34	4.5 4.51	1.0 1.05	10.0 10.18	10.0 9.83	0.03 0.04	0.02 0.023	0.13 0.13	Bal. -
I-26	2	8.0 7.81	10.0 10.10	4.5 4.38	6.3 6.33	1.0 1.03	10.0 10.01	10.0 9.78	0.03 0.03	0.02 0.020	0.13 0.13	Bal. -
I-27	3	8.0 7.84	10.0 10.26	4.5 4.26	6.3 6.53	1.8 1.98	10.0 10.14	10.0 9.71	0.03 0.04	0.02 0.021	0.13 0.13	Bal. -

* Aim is shown first. All data is in weight percent.
Trace Elements: Fe <0.30, Mn <0.05, Si <0.15, and S <0.01.

TABLE 15

Stress Rupture Data for Series I Alloys

Tested at 2000°F and 15,000 psi

Alloy	Life (hrs)	$\log_{10}$ (Life x 10^2)	Elongation (%)	R.A. (%)
I-1	0.01*	0.0000	73.1	92.4
	0.01*	0.0000	72.6	95.6
	0.01*	0.0000	73.5	86.2
I-2	0.07	0.8451	41.4	42.2
	0.06	0.7782	40.3	34.6
	0.08	0.9031	38.6	32.8
I-3	0.10	1.000	30.2	35.0
	0.13	1.1139	28.0	35.0
	0.10	1.0000	44.7	36.4
I-4	0.24	1.3802	23.3	24.1
	0.30	1.4771	23.8	25.0
	0.21	1.3222	28.8	35.8
I-5	9.4	2.9736	9.7	9.0
	12.4	3.0934	7.9	7.0
	13.8	3.1399	8.1	8.5
I-6	5.6	2.7482	3.5	2.0
	12.9	3.1106	3.3	1.2
	5.8	2.7634	1.8	2.0
	5.4**		1.1	2.0
	9.2**		0.8	1.2
I-7	14.0	3.1461	6.6	13.0
	16.7	3.2227	2.6	7.0
	18.5	3.2672	7.4	8.1
I-8	5.4	2.7324	2.5	3.1
	2.2	2.3424	3.3	3.9
	6.2	2.7924	2.8	3.6
I-9	2.0	2.3010	3.4	3.1
	2.6	2.4150	7.0	2.7
	0.6	1.7782	6.5	4.7
I-10	0.30	1.4771	29.8	36.2
	0.24	1.3802	23.1	37.5
	0.18	1.2553	33.1	31.6

* Broke on loading. Assigned arbitrary value of 0.01 hrs.

** Additional tests were conducted in an attempt to verify previous results. These data were not used in the Latin Square calculations.

TABLE 15 (Continued)

Alloy	Life (hrs)	Log_{10} (Life x 10^2)	Elongation (%)	R.A. (%)
I-11	4.7	2.6721	1.4	3.5
	3.9	2.5911	4.2	7.7
	3.7	2.5682	7.6	8.5
I-12	2.1	2.3222	16.9	22.4
	2.5	2.3979	11.8	15.5
	2.7	2.4314	17.2	23.1
I-13	8.3	2.9191	4.0	6.6
	9.5	2.9777	6.1	6.2
	7.6	2.8808	4.9	7.4
I-14	4.7	2.6721	4.4	2.0
	4.2	2.6232	2.3	2.4
	3.5	2.5441	2.5	2.8
I-15	3.0	2.4771	4.7	3.2
	2.2	2.3424	3.8	3.6
	2.3	2.3617	5.4	4.7
I-16	0.9	1.9542	26.1	39.4
	1.1	2.0414	25.9	38.1
	1.0	2.0000	25.9	47.9
I-17	2.2	2.3424	6.7	8.6
	2.2	2.3424	5.9	7.7
	2.8	2.4472	7.4	8.6
I-18	5.3	2.7243	7.0	4.0
	7.8	2.8921	4.5	8.2
	4.9	2.6902	7.6	7.0
I-19	5.3	2.7243	5.3	7.4
	6.1	2.7853	3.7	5.1
	5.7	2.7559	6.9	5.5
I-20	2.5	2.3979	2.7	4.5
	2.3	2.3617	4.1	4.7
	2.8	2.4472	5.1	5.1
I-21	2.0	2.3010	4.2	4.3
	2.5	2.3979	4.7	4.9
	2.3	2.3617	4.2	3.9
I-22	0.21	1.3222	25.5	54.9
	0.32	1.5051	40.3	61.5
	0.24	1.3802	29.2	62.8
I-23	1.3	2.1139	9.1	14.9
	1.5	2.1761	9.0	12.0
	1.2	2.0792	10.3	15.2
I-24	2.7	2.4314	6.4	10.1
	2.6	2.4150	5.6	7.8
	2.6	2.4150	7.1	7.8

TABLE 15 (Continued)

Alloy	Life (hours)	Log_{10} ($Life \times 10^2$)	Elongation (%)	R.A. (%)
I-25	1.4	2.1461	9.4	14.6
	1.5	2.1761	9.3	16.2
	1.8	2.2553	12.2	15.7
I-26	0.35	1.5441	14.9	19.2
	0.35	1.5441	16.3	23.1
	0.42	1.6232	14.8	20.5
I-27	0.25	1.3979	12.6	17.4
	0.24	1.3802	10.4	13.5
	0.20	1.3010	14.6	21.0

TABLE 16

Room Temperature Tensile Results for Series I Alloys

Alloy	Ultimate (10^3 psi)	0.2% Offset Yield(10^3psi)	Elongation (%)	R.A. (%)
I-1	104.7 105.2	85.4 84.5	4.1 5.6	5.9 10.2
I-2	106.3 95.9	88.6 81.4	8.0 10.4	15.6 18.8
I-3	123.1 117.2	94.2 93.3	9.3 8.0	12.7 12.3
I-4	128.1 141.8	102.8 113.0	5.6 8.0	8.5 13.1
I-5	135.1 133.6	133.4 114.0	7.1 6.8	7.1 7.0
I-6	137.8 144.4	120.7 111.0	4.6 10.1	6.3 9.2
I-7	148.2 148.2	130.1 -	3.9 5.0	7.4 8.9
I-8	149.4 144.4	143.1 143.2	0.2 0.3	1.6 1.2
I-9	135.4 145.2	- -	0.3 0.2	1.2 0.8
I-10	120.6 125.3	97.9 99.5	6.2 6.9	10.0 9.3
I-11	124.9 123.1	109.5 106.7	5.1 6.5	6.9 10.8
I-12	134.1 137.1	109.9 109.1	7.7 9.3	9.7 11.0
I-13	143.9 142.7	127.9 126.5	3.4 3.7	7.4 5.9
I-14	150.5 153.2	- 151.8	0.6 0.4	1.6 1.6
I-15	137.6 138.6	- -	0.0 0.3	0.0 0.4
I-16	125.9 124.5	119.9 119.0	1.9 2.2	4.2 2.7

TABLE 16 (Continued)

Alloy	Ultimate (10^3 psi)	0.2% Offset Yield(10^3psi)	Elongation (%)	R.A. (%)
I-17	132.5	129.7	2.1	2.0
	132.4	129.0	1.8	1.6
I-18	146.7	146.3	0.7	1.2
	150.2	147.8	1.4	2.4
I-19	142.1	127.3	2.5	5.8
	143.9	126.3	3.9	7.8
I-20	151.0	-	1.2	1.6
	149.0	147.8	1.0	1.2
I-21	148.6	-	0.0	0.0
	147.3	-	0.6	0.0
I-22	116.7	116.1	1.4	1.6
	116.7	-	1.8	6.3
I-23	129.7	125.5	1.7	3.6
	133.3	128.2	1.9	2.4
I-24	156.2	-	0.5	0.0
	152.4	-	0.8	0.0
I-25	152.4	150.8	0.3	0.0
	152.2	151.8	0.8	0.8
I-26	101.2	-	0.0	0.0
	102.9	-	0.0	0.0
I-27	89.7	-	1.6	0.9
	94.0	-	0.6	0.4

TABLE 17

1400°F Tensile Results for Series I Alloys

Alloy	Ultimate (10^3 psi)	0.2% Offset Yield (10^3 psi)	Elongation (%)	R.A. (%)
I-1	92.0	65.7	2.9	0.8
	88.1	72.2	1.8	2.4
I-2	84.5	79.7	3.0	5.1
	99.8	78.0	2.0	4.3
I-3	116.3	87.8	4.6	7.0
	114.0	88.1	3.3	8.6
I-4	106.0	85.3	2.8	7.7
	98.7	85.5	2.8	5.5
I-5	139.7	114.4	4.2	4.7
	136.8	89.8	3.1	4.0
I-6	142.4	120.0	2.6	3.9
	140.4	117.0	3.0	6.2
I-7	147.4	115.7	4.8	7.3
	148.2	118.2	2.8	4.7
I-8	145.6	-	0.6	0.8
	139.2	137.9	1.0	0.8
I-9	135.3	-	0.3	0.0
	143.2	137.1	0.9	0.0
I-10	111.0	83.7	5.2	8.1
	108.8	86.4	3.5	7.5
I-11	132.9	110.3	3.9	4.7
	130.0	110.8	3.3	5.1
I-12	120.0	95.1	3.0	6.2
	130.5	97.5	5.7	5.5
I-13	142.4	116.6	2.7	3.5
	136.4	115.9	1.9	4.9
I-14	135.7	-	0.4	0.8
	129.5	-	0.6	0.0
I-15	128.7	-	0.2	0.0
	136.1	-	0.2	0.0
I-16	116.1	107.8	2.2	1.6
	115.8	107.1	1.8	4.3

TABLE 17 (Continued)

Alloy	Ultimate (10^3 psi)	0.2% Offset Yield (10^3 psi)	Elongation (%)	R.A. (%)
I-17	127.1	-	0.4	1.2
	119.6	-	0.4	1.6
I-18	146.2	139.0	0.3	1.2
	144.2	138.9	0.8	0.8
I-19	145.0	118.8	2.9	3.9
	145.4	120.2	1.8	3.9
I-20	141.9	-	0.8	0.0
	146.4	142.0	1.1	0.8
I-21	144.9	-	0.9	0.0
	140.2	-	0.3	0.0
I-22	94.5	-	2.1	0.4
	92.8	-	3.3	0.8
I-23	130.8	129.3	1.2	0.8
	122.4	-	1.4	1.2
I-24	141.2	139.6	1.3	0.4
	129.5	-	0.9	0.8
I-25	137.1	-	0.9	0.4
	138.1	-	1.2	0.4
I-26	93.5	-	0.7	0.0
	93.0	-	0.6	0.0
I-27	98.0	-	0.6	0.0
	78.2	-	0.5	0.0

TABLE 18

1875°F Tensile Results for Series I Alloys

Alloy	Ultimate (10^3psi)	0.2% Offset Yield(10^3psi)	Elongation (%)	R.A. (%)
I-1	15.0	14.0	31.0	33.8
	15.5	14.3	22.7	34.2
I-2	25.5	20.0	-	19.6
	25.7	19.8	22.3	22.3
I-3	26.2	20.6	19.5	16.0
	25.7	21.3	25.0	25.9
I-4	29.5	23.1	18.7	18.0
	29.7	23.7	21.4	18.8
I-5	51.7	40.5	12.8	11.2
	49.9	44.1	12.4	12.7
I-6	56.9	49.0	3.8	3.5
	55.1	47.9	4.5	4.7
I-7	54.1	42.4	9.9	11.5
	56.0	47.1	8.5	10.4
I-8	51.2	46.9	6.5	7.0
	50.0	46.0	5.1	6.2
I-9	46.4	42.2	5.1	6.2
	47.2	44.0	4.0	5.5
I-10	29.2	22.3	21.3	28.4
	29.5	23.5	23.9	25.6
I-11	46.5	40.5	7.1	8.5
	47.7	41.8	7.4	10.4
I-12	39.4	31.2	16.1	21.0
	39.5	32.5	16.5	18.8
I-13	50.1	43.5	8.1	9.3
	51.0	44.4	5.8	6.8
I-14	49.2	43.8	7.1	10.0
	48.5	43.7	7.3	11.2
I-15	45.3	41.8	7.9	10.4
	44.5	40.4	7.7	8.6
I-16	42.1	34.2	13.5	16.6
	43.9	37.0	11.9	11.1
I-17	44.5	38.8	7.2	8.5
	45.2	39.9	9.7	15.0

TABLE 18 (Continued)

Alloy	Ultimate (10^3psi)	0.2% Offset Yield(10^3psi)	Elongation (%)	R.A. (%)
I-18	45.9	41.6	9.4	12.6
	47.5	43.0	8.9	13.8
I-19	44.3	36.4	11.7	13.3
	46.7	38.6	8.9	10.0
I-20	43.7	37.7	7.8	12.8
	46.6	41.5	5.9	12.2
I-21	45.5	41.2	6.8	14.2
	49.5	44.6	7.3	12.8
I-22	36.2	29.3	14.8	26.7
	37.0	31.1	13.3	22.3
I-23	43.2	37.5	15.4	19.9
	42.9	38.0	11.9	16.8
I-24	46.3	41.4	11.7	17.3
	43.5	38.1	10.2	14.4
I-25	44.3	38.4	15.0	22.5
	41.7	36.0	17.5	20.3
I-26	36.4	31.9	12.4	13.1
	34.8	30.4	12.9	15.3
I-27	34.7	30.6	10.3	14.7
	33.2	29.0	9.2	11.9

TABLE 19

2000°F Tensile Results for Series I Alloys

Alloy	Ultimate (10^3 psi)	0.2% Offset Yield(10^3psi)	Elongation (%)	R.A. (%)
I-1	5.0	4.6	72.6	94.4
	5.0	4.4	71.8	92.4
I-2	11.1	9.3	42.1	46.3
	11.1	9.3	32.5	40.1
I-3	11.3	9.8	35.6	31.1
	11.9	9.9	42.9	47.6
I-4	16.0	12.8	30.3	22.0
	15.6	13.0	26.4	23.7
I-5	32.5	29.1	17.5	13.6
	31.3	27.1	13.2	10.4
I-6	34.4	30.9	5.2	5.1
	35.5	29.9	3.7	3.1
I-7	36.1	30.5	10.0	13.3
	34.7	29.9	15.7	15.1
I-8	32.7	29.3	5.5	8.2
	32.6	29.4	7.2	8.9
I-9	31.6	28.6	10.7	11.8
	31.4	29.5	6.9	5.5
I-10	14.3	11.8	27.1	30.1
	14.3	12.0	28.2	30.7
I-11	28.3	24.5	6.0	9.3
	28.0	24.1	12.0	8.5
I-12	24.0	19.2	15.4	26.9
	23.6	19.3	22.8	23.2
I-13	30.5	25.9	11.3	11.5
	29.5	25.0	10.5	8.8
I-14	30.6	27.4	10.6	11.5
	32.3	29.0	8.6	9.3
I-15	29.0	26.2	5.7	8.9
	28.1	25.8	8.3	9.7
I-16	23.0	18.9	22.2	31.4
	23.1	18.7	16.5	33.7
I-17	26.1	22.6	10.7	15.7
	24.7	21.5	16.2	15.6

TABLE 19 (Continued)

Alloy	Ultimate (10^3 psi)	0.2% Offset Yield(10^3psi)	Elongation (%)	R.A. (%)
I-18	29.4	25.8	13.2	14.4
	24.3	21.6	11.4	34.6
I-19	28.3	23.5	13.1	15.9
	30.4	25.8	13.8	15.3
I-20	30.6	27.2	8.1	13.9
	30.5	26.6	10.8	10.4
I-21	30.3	27.4	8.6	11.0
	31.1	27.8	10.7	8.9
I-22	17.8	14.9	26.5	44.1
	20.0	15.1	20.6	50.3
I-23	24.8	23.4	15.6	24.4
	26.0	21.5	11.3	16.8
I-24	30.2	27.3	13.0	12.9
	28.9	25.5	12.3	17.0
I-25	26.4	22.4	15.5	24.7
	25.2	21.2	10.9	22.6
I-26	21.5	18.9	16.4	22.7
	21.4	18.6	15.1	15.6
I-27	19.8	16.2	17.3	19.1
	19.8	17.5	18.5	19.6

TABLE 20

Aim and Actual Chemical Compositions for Series II Alloys *

Alloy	Components Varied at Three Levels			Fixed Components										
	Cb	V	Hf	Co	Cr	Zr	B	C	Ni	Mo	W	Ta	Ti	Al
IIa	0	0	0	10.0	10.0	0.03	0.02	0.13	Bal.	2.0	5.5	8.0	1.0	4.5
	-	-	-	10.0	9.81	0.03	0.023	0.11	-	1.95	5.38	7.80	1.02	4.58
IIb	0	1.0	1.0	10.0	10.0	0.03	0.02	0.13	Bal.	2.0	5.5	8.0	1.0	4.5
	-	0.97	0.75	10.05	9.80	0.05	0.02	0.12	-	1.97	5.49	7.84	1.02	4.52
IIc	0	2.0	2.0	10.0	10.0	0.03	0.02	0.13	Bal.	2.0	5.5	8.0	1.0	4.5
	-	1.96	1.75	10.01	9.80	0.07	0.019	0.12	-	1.97	5.39	7.74	1.03	4.55
IId	1.0	0	1.0	10.0	10.0	0.03	0.02	0.13	Bal.	2.0	5.5	8.0	1.0	4.5
	0.98	-	0.98	10.02	9.84	0.07	0.018	0.12	-	2.00	5.48	7.80	0.98	4.50
IIe	1.0	1.0	2.0	10.0	10.0	0.03	0.02	0.13	Bal.	2.0	5.5	8.0	1.0	4.5
	1.02	0.96	1.78	10.02	9.83	0.08	0.019	0.13	-	1.98	5.44	7.78	0.99	4.52
IIf	1.0	2.0	0	10.0	10.0	0.03	0.02	0.13	Bal.	2.0	5.5	8.0	1.0	4.5
	1.05	1.94	-	10.00	9.79	0.04	0.019	0.12	-	1.92	5.37	7.79	1.03	4.49
IIg	2.0	0	2.0	10.0	10.0	0.03	0.02	0.13	Bal.	2.0	5.5	8.0	1.0	4.5
	1.90	-	2.00	10.03	9.80	0.07	0.017	0.12	-	2.10	5.45	7.81	0.99	4.58
IIh	2.0	1.0	0	10.0	10.0	0.03	0.02	0.13	Bal.	2.0	5.5	8.0	1.0	4.5
	2.00	0.98	-	10.00	9.85	0.04	0.017	0.12	-	1.96	5.50	7.79	1.02	4.52
IIi	2.0	2.0	1.0	10.0	10.0	0.03	0.02	0.13	Bal.	2.0	5.5	8.0	1.0	4.5
	2.01	1.96	0.82	10.08	9.78	0.07	0.023	0.13	-	2.00	5.39	7.81	1.04	4.52

* Aim is shown first. All data is in weight percent.
Trace Elements: Fe < 0.30, Mn < 0.05, Si < 0.15, and S < 0.01.

TABLE 21

Stress Rupture Data for Series II Alloys

Tested at 2000°F and 15,000 psi

Alloy	Life (hrs)	Log_{10} (Life)	Elongation (%)	R. A. (%)
IIa	7.44	0.8716	7.8	5.7
	9.9	0.9956	3.8	6.9
	6.5	0.8129	6.9	7.0
IIb	15.02	1.1767	19.0	22.4
	14.47	1.1605	12.1	21.7
	13.39	1.1268	18.6	19.9
IIc	12.59	1.1000	7.7	13.5
	13.3	1.1239	6.0	17.0
	13.59	1.1332	9.1	17.4
IId	18.7	1.2718	13.1	16.3
	17.3	1.2380	9.9	16.7
	15.4	1.1875	13.5	18.8
IIe	12.8	1.1072	11.2	20.3
	13.26	1.1225	7.7	12.7
	14.9	1.1732	5.0	12.2
IIf	6.6	0.8195	2.2	2.0
	4.66	0.6684	4.9	6.2
	3.8	0.5798	3.4	3.9
IIg	14.79	1.1700	4.7	11.4
	26.5	1.4232	9.2	12.6
	16.39	1.2146	2.9	10.8
IIh	10.5	1.0212	2.5	2.0
	9.4	0.9731	1.2	2.0
	6.49	0.8122	3.1	3.1
IIi	7.04	0.8476	9.1	8.9
	7.6	0.8808	9.8	14.5
	6.33	0.8014	7.9	11.6

TABLE 22

Room Temperature Tensile Results for Series II Alloys

Alloy	Ultimate (10^3psi)	0.2% Offset Yield (10^3psi)	Elongation (%)	R. A. (%)
IIa	145.9 147.0	124.9 123.9	6.2 6.6	7.8 6.6
IIb	155.0 151.6	135.3 132.5	5.6 6.0	8.1 7.0
IIc	149.2 154.6	135.3 142.0	3.2 4.6	8.1 4.7
IId	151.0 151.4	132.6 130.6	2.8 3.7	7.1 7.0
IIe	153.6 149.0	141.3 139.2	0.5 1.7	5.1 2.7
IIf	161.0 151.8	139.3 134.9	5.3 3.7	6.7 6.7
IIg	147.3 152.2	138.4 144.2	1.6 1.4	4.0 3.2
IIh	167.3 164.2	142.7 144.1	6.4 5.0	6.3 6.3
IIi	153.8 151.6	148.8 145.2	1.8 0.7	1.6 2.0

TABLE 23

$1400^{o}F$ Tensile Results for Series II Alloys

Alloy	Ultimate (10^3 psi)	0.2% Offset Yield (10^3 psi)	Elongation (%)	R. A. (%)
IIa	135.2	104.9	2.8	5.9
	133.5	109.1	3.3	4.7
IIb	147.8	116.1	5.1	8.5
	147.0	116.4	5.0	9.6
IIc	152.4	131.3	5.0	4.7
	156.8	133.9	3.0	5.5
IId	150.0	119.8	4.5	6.6
	150.2	116.9	4.8	6.3
IIe	154.0	123.4	5.0	5.9
	157.2	129.6	5.9	5.9
IIf	152.2	128.7	2.2	4.7
	152.2	129.9	1.8	3.1
IIg	158.0	136.3	3.2	6.6
	159.5	141.4	4.1	5.1
IIh	156.6	134.1	3.8	5.5
	154.2	130.9	3.4	6.2
IIi	156.6	143.7	2.2	2.0
	160.2	147.4	2.0	1.6

TABLE 24

1875°F Tensile Results for Series II Alloys

Alloy	Ultimate (10^3 psi)	0.2% Offset Yield (10^3 psi)	Elongation (%)	R. A. (%)
IIa	48.7 50.3	40.1 40.1	13.3 6.5	10.8 7.8
IIb	46.4 45.8	37.4 37.9	12.7 14.8	14.5 16.3
IIc	54.5 53.0	47.8 45.6	12.1 15.2	14.5 15.6
IId	54.9 53.0	45.6 43.8	13.8 14.4	16.3 14.1
IIe	54.9 53.8	47.3 44.9	8.0 11.1	13.1 11.5
IIf	49.2 47.8	42.1 41.1	6.2 6.4	9.3 7.0
IIg	59.6 56.9	51.0 48.2	10.8 7.1	12.7 12.3
IIh	53.1 53.7	44.8 43.8	8.3 8.2	8.5 9.3
IIi	45.3 45.7	36.6 39.1	4.9 8.4	9.3 11.2

TABLE 25

2000°F Tensile Results for Series II Alloys

Alloy	Ultimate (10^3 psi)	0.2% Offset Yield(10^3psi)	Elongation (%)	R. A. (%)
IIa	31.1	25.8	8.6	13.1
	31.7	26.1	9.9	11.9
IIb	33.1	27.9	17.6	18.5
	31.7	26.9	15.7	23.1
IIc	33.3	29.3	15.5	15.3
	33.4	29.2	11.9	17.0
IId	34.6	29.8	13.8	18.9
	34.6	29.3	17.3	21.0
IIe	36.8	31.8	7.7	10.0
	34.8	30.0	8.7	13.1
IIf	31.9	27.0	8.9	12.7
	31.5	27.7	12.6	8.9
IIg	37.9	33.2	9.7	10.8
	39.4	33.9	7.7	9.3
IIh	35.4	30.3	9.1	8.5
	37.0	31.3	7.7	10.1
IIi	31.3	27.1	10.7	13.4
	31.8	27.8	9.6	11.9

TABLE 26

Aim and Actual Chemical Compositions for Series III Alloys*

Alloy	Components Varied at Three Levels			Fixed Components									
	Al	Ru	Re	Co	Cr	Zr	B	C	Ni	Mo	W	Ta	Ti
IIIa	6.3 6.15	0 -	0 -	10.0 9.75	10.0 9.51	0.03 0.028	0.02 0.020	0.13 0.12	Bal. -	2.0 2.17	5.5 5.2	8.0 8.2	1.0 0.87
IIIb	5.4 5.32	1.0 0.94	0 -	10.0 9.66	10.0 9.61	0.03 0.025	0.02 0.019	0.13 0.11	Bal. -	2.0 2.18	5.5 5.1	8.0 8.1	1.0 0.94
IIIc	4.5 4.64	2.0 2.25	0 -	10.0 9.31	10.0 8.97	0.03 0.037	0.02 0.018	0.13 0.11	Bal. -	2.0 2.12	5.5 5.1	8.0 8.1	1.0 0.85
IIId	5.4 5.33	0 -	2.0 1.45	10.0 9.32	10.0 9.26	0.03 0.049	0.02 0.024	0.13 0.10	Bal. -	2.0 2.12	5.5 5.2	8.0 8.5	1.0 0.90
IIIe	4.5 4.53	1.0 1.50	2.0 1.41	10.0 9.42	10.0 9.24	0.03 0.042	0.02 0.021	0.13 0.11	Bal. -	2.0 2.06	5.5 5.2	8.0 8.2	1.0 0.87
IIIf	6.3 6.27	2.0 2.35	2.0 1.56	10.0 9.27	10.0 9.08	0.03 0.040	0.02 0.024	0.13 0.13	Bal. -	2.0 2.12	5.5 4.9	8.0 8.4	1.0 0.86
IIIg	4.5 4.40	0 -	4.0 2.80	10.0 9.22	10.0 9.00	0.03 0.039	0.02 0.023	0.13 0.11	Bal. -	2.0 1.99	5.5 5.1	8.0 8.1	1.0 0.85
IIIh	6.3 6.26	1.0 1.80	4.0 3.16	10.0 9.53	10.0 8.82	0.03 0.045	0.02 0.018	0.13 0.10	Bal. -	2.0 2.06	5.5 5.3	8.0 8.5	1.0 0.82
IIIi	5.4 5.48	2.0 2.85	4.0 2.99	10.0 9.81	10.0 8.77	0.03 0.047	0.02 0.021	0.13 0.11	Bal. -	2.0 2.13	5.5 5.3	8.0 8.5	1.0 0.83

* Aim is shown first. All data is in weight percent.
Trace Elements: Fe < 0.30, Mn < 0.05, Si < 0.15, and S < 0.01.

TABLE 27

Stress Rupture Data for Series III Alloys

Tested at 2000°F and 15,000 psi

Alloy	Life (hrs.)	Log_{10} (Life)	Elongation (%)	R. A. (%)
IIIa	7.58	0.8797	3.3	2.0
	6.64	0.8222	2.3	1.2
	7.6	0.8808	1.8	1.2
IIIb	9.7	0.9868	1.8	1.6
	8.1	0.9085	1.3	1.6
	8.32	0.9201	2.5	2.4
IIIc	8.12	0.9096	3.7	4.3
	9.0	0.9542	3.2	3.9
	8.61	0.9350	1.8	2.4
IIId	15.11	1.1793	5.0*	1.2*
	18.9	1.2770	1.6	1.2
	15.9	1.2014	0.8	2.4
IIIe	10.0	1.0000	2.4	0.4
	10.92	1.0382	2.6	1.2
	13.0	1.1139	3.9	2.4
IIIf	5.36	0.7292	2.1	2.0
	4.38	0.6415	N.A.	N.A.
	5.3	0.7243	2.4	1.6
IIIg	15.5	1.1903	2.1	0.8
	17.0	1.2304	2.4	1.6
	14.08	1.1485	3.3	2.4
III h	3.9	0.5911	2.2	1.6
	4.15	0.6180	2.3	2.0
	4.4	0.6435	2.0	1.6
III i	12.6	1.1004	1.7	0.8
	11.4	1.0569	0.8	1.2
	10.2	1.0086	1.6	0.8

* Broke into 3 pieces

TABLE 28

Room Temperature Tensile Results for Series III Alloys

Alloy	Ultimate (10^3 psi)	0.2% Offset Yield(10^3 psi)	Elongation (%)	R. A. (%)
IIIa	145.5 147.1	127.7 131.2	4.1 3.6	4.8 4.3
IIIb	157.0 149.0	132.7 128.5	4.4 4.8	7.8 6.7
IIIc	151.2 150.8	128.3 129.1	6.7 6.5	8.9 7.8
IIId	155.4 155.0	137.1 131.5	4.6 5.7	6.3 8.4
IIIe	157.6 158.4	132.3 131.5	6.0 6.5	10.1 7.4
IIIf	153.0 152.2	146.8 147.8	1.2 1.6	1.2 1.6
IIIg	149.4 161.4	131.7 136.3	4.8 6.5	8.5 7.0
IIIh	145.6 150.4	- -	0.16 0.2	0.0 0.8
IIIi	162.4 165.9	151.0 152.8	1.7 2.8	3.1 2.0

TABLE 29

1400°F Tensile Results for Series III Alloys

Alloy	Ultimate (10^3 psi)	0.2% Offset Yield(10^3 psi)	Elongation (%)	R. A. (%)
IIIa	151.4 150.6	134.3 131.3	2.2 3.3	3.5 3.5
IIIb	152.2 151.8	124.4 122.9	2.9 3.9	5.9 3.9
IIIc	139.8 142.0	109.6 113.2	3.7 4.3	7.0 6.6
IIId	159.6 154.6	131.2 129.5	3.3 2.5	3.5 3.3
IIIe	149.2 144.4	117.8 115.2	3.2 2.5	6.6 6.2
IIIf	140.1 144.9	None 140.7	1.2 1.3	0.8 0.8
IIIg	151.2 148.4	118.1 119.7	4.3 3.0	5.5 5.9
IIIh	139.5 122.1*	- -	0.6 0.1	0.8 0.0
IIIi	157.6 160.6	144.0 144.3	1.5 1.1	2.0 2.7

* Broke in threads

TABLE 30

1875°F Tensile Results for Series III Alloys

Alloy	Ultimate (10^3 psi)	0.2% Offset Yield (10^3 psi)	Elongation (%)	R. A. (%)
III a	54.7 53.6	46.7 44.9	5.7 3.4	4.3 3.9
III b	53.1 52.1	43.1 38.5	2.5 3.3	3.9 6.6
III c	57.8 56.2	45.8 45.1	7.2 5.8	6.3 7.4
III d	64.8 65.5	58.2 57.9	4.9 5.1	6.2 6.3
III e	55.3 53.4	44.9 42.7	6.0 7.0	4.7 6.3
IIIf	55.2 54.9	47.5 48.9	3.7 4.1	4.7 3.9
IIIg	63.9 60.4	52.6 49.2	6.8 8.6	7.8 8.5
IIIh	57.7 57.3	53.4 51.8	2.5 2.2	2.0 2.7
IIIi	73.5 74.0	67.5 67.0	2.5 3.1	3.2 2.8

TABLE 31

2000°F Tensile Results for Series III Alloys

Alloy	Ultimate (10^3psi)	0.2% Offset Yield(10^3psi)	Elongation (%)	R. A. (%)
IIIa	41.1 39.9	36.5 35.2	5.0 5.6	5.5 5.9
IIIb	39.0 39.0	33.8 32.2	3.4 2.9	5.5 3.9
IIIc	33.0 32.5	28.3 27.5	10.4 9.5	10.1 11.3
IIId	39.3 43.0	35.2 36.1	6.0 6.1	7.8 7.0
III e	39.3 36.4	32.0 31.2	6.4 7.2	6.3 7.0
III f	36.7 36.5	32.7 35.7	5.5 6.1	5.9 7.0
III g	38.6 38.0	32.6 32.5	6.8 9.4	8.9 9.7
III h	37.4 38.6	35.2 35.8	6.0 5.3	5.5 4.7
III i	46.7 47.3	37.9* 40.1*	5.2 4.7	6.3 4.7

* 0.1% offset yield

TABLE 32

Aim and Actual Chemical Components for Series IV Alloys *

Alloy	Components Varied at Two Levels								Fixed Components					
	Co	Cr	Ta	W	Hf	Re	Zr	C	Al	Mo	Ti	Cb	B**	Ni
IV A	10.0	7.0	8.0	7.5	1.5	0	0.03	0.08	5.4	2.0	1.0	0.5	0.02	Bal.
	9.89	6.95	7.6	7.50	1.50	-	0.05	0.084	5.56	2.02	0.98	0.56	0.02	-
IV B	10.0	10.0	10.0	7.5	2.5	1.0	0.03	0.08	5.4	2.0	1.0	0.5	0.02	Bal.
	9.95	9.28	10.0	7.48	2.50	0.88	0.06	0.076	5.50	1.98	1.00	0.55	-	-
IV C	5.0	10.0	8.0	7.5	1.5	1.0	0.23	0.08	5.4	2.0	1.0	0.5	0.02	Bal.
	5.0	9.83	7.9	7.44	1.48	0.86	0.25	0.068	5.58	2.00	1.00	0.57	-	-
IV D	10.0	7.0	10.0	5.5	1.5	1.0	0.23	0.08	5.4	2.0	1.0	0.5	0.02	Bal.
	9.88	6.78	9.5	5.51	1.68	0.90	0.22	0.076	5.47	2.01	0.96	0.56	-	-
IV E	5.0	10.0	10.0	5.5	2.5	1.0	0.23	0.18	5.4	2.0	1.0	0.5	0.02	Bal.
	5.01	9.70	10.0	5.46	2.36	0.88	0.24	0.16	5.50	1.99	0.99	0.58	-	-
IV F	5.0	10.0	8.0	7.5	2.5	0	0.03	0.18	5.4	2.0	1.0	0.5	0.02	Bal.
	4.98	9.51	9.2	7.45	2.41	-	0.06	0.17	5.52	2.00	1.02	0.58	0.023	-
IV G	5.0	10.0	10.0	5.5	1.5	0	0.03	0.08	5.4	2.0	1.0	0.5	0.02	Bal.
	4.96	9.90	9.5	5.48	1.43	-	0.05	0.088	5.50	1.98	1.0	0.57	0.018	-
IV H	10.0	7.0	10.0	5.5	2.5	0	0.03	0.18	5.4	2.0	1.0	0.5	0.02	Bal.
	9.89	6.70	9.9	5.51	2.59	-	0.06	0.17	5.47	1.97	1.01	0.59	0.023	-
IV I	5.0	7.0	10.0	7.5	2.5	0	0.23	0.08	5.4	2.0	1.0	0.5	0.02	Bal.
	4.98	6.62	10.1	7.51	2.30	-	0.22	0.078	5.51	1.98	0.98	0.58	0.019	-

* Aim is shown first. All data is in weight percent. Trace Elements: Fe $<$ 0.30, Mn $<$ 0.05, Si $<$ 0.15, and S $<$ 0.01.

** It appears that Re interferred with the B determination by wet chemical method used to analyze this series. As a result, no B result is reported for those alloys containing Re. In the other series where B and Re were present, B was determined using a dual grating emission spectrograph.

TABLE 32 (Continued)

Aim and Actual Chemical Components for Series IV Alloys *

Alloy	Components Varied at Two Levels								Fixed Components					
	Co	Cr	Ta	W	Hf	Re	Zr	C	Al	Mo	Ti	Cb	B**	Ni
IV J	10.0	7.0	8.0	7.5	2.5	1.0	0.23	0.18	5.4	2.0	1.0	0.5	0.02	Bal.
	9.96	6.85	7.8	7.53	2.36	0.87	0.22	0.16	5.49	2.00	0.96	0.58	-	-
IV K	10.0	10.0	8.0	5.5	2.5	0	0.23	0.08	5.4	2.0	1.0	0.5	0.02	Bal.
	9.94	10.20	7.8	5.50	2.52	-	0.22	0.084	5.50	2.08	1.01	0.58	0.019	-
IV L	5.0	7.0	8.0	5.5	1.5	0	0.23	0.18	5.4	2.0	1.0	0.5	0.02	Bal.
	4.89	7.21	7.8	5.50	1.53	-	0.21	0.17	5.52	2.08	1.02	0.57	0.022	-
IV M	5.0	7.0	10.0	7.5	1.5	1.0	0.03	0.18	5.4	2.0	1.0	0.5	0.02	Bal.
	4.98	6.53	10.0	7.50	1.45	0.88	0.05	0.18	5.47	1.96	1.01	0.57	-	-
IV N	5.0	7.0	8.0	5.5	2.5	1.0	0.03	0.08	5.4	2.0	1.0	0.5	0.02	Bal.
	4.98	6.90	8.1	5.50	2.14	0.87	0.06	0.082	5.49	2.00	1.00	0.51	-	-
IV O	10.0	10.0	8.0	5.5	1.5	1.0	0.03	0.18	5.4	2.0	1.0	0.5	0.02	Bal.
	9.94	10.10	8.0	5.50	1.43	0.86	0.05	0.17	5.50	2.02	0.99	0.53	-	-
IV P	10.0	10.0	10.0	7.5	1.5	0	0.23	0.18	5.4	2.0	1.0	0.5	0.02	Bal.
	9.83	9.79	9.7	7.46	1.75	-	0.23	0.15	5.50	2.06	1.01	0.54	0.022	-
IV Y	5.0	6.0	8.0	5.5	2.0	1.0	0.03	0.15	5.4	2.0	1.0	1.0	0.02	Bal.
	4.98	5.89	7.9	5.54	1.62	0.80	0.06	0.13	5.37	2.09	1.01	1.04	-	-
IV Z	7.5	8.5	9.0	6.5	2.0	1.0	0.13	0.13	5.4	2.0	1.0	0.5	0.02	Bal.
	7.48	8.37	9.0	6.56	1.41	1.02	0.13	0.12	5.46	2.07	1.02	0.58	-	-

* Aim is shown first. All data is in weight percent. Trace Elements: Fe $<$ 0.30, Mn $<$ 0.05, Si $<$ 0.15, and S $<$ 0.01.

** It appears that Re interferred with the B determination by wet chemical method used to analyze this series. As a result, no B result is reported for those alloys containing Re. In the other series where B and Re were present, B was determined using a dual grating emission spectrograph.

TABLE 33 (Continued)

Alloy	Life (hours)	Log 10 (Life)	Elongation (%)	R.A. (%)
IV M	22.0	1.3424	2.3	1.6
	25.7	1.4099	5.8*	2.4
	23.3	1.3674	2.7	2.7
IV N	38.7	1.5877	4.5	2.0
	31.9	1.5038	1.5	3.0
	29.0	1.4624	1.8	2.7
IV O	14.0	1.1461	7.9	12.3
	16.3	1.2122	7.5	8.9
	17.8	1.2504	6.4	7.8
IV P	5.0	0.6990	3.0	9.3
	5.1	0.7076	3.4	6.6
	6.6	0.8195	N.A.	N.A.
IV Y	42.39	1.6272	2.6	2.3
	45.3	1.6561	2.9	3.9
	37.1	1.5694	5.7	7.7
IV Z	18.5	1.2672	5.6	11.5
	12.5	1.0969	4.7	3.9
	14.3	1.1553	5.0	10.0

* 4 piece fracture

TABLE 33

Stress Rupture Data for Series IV Alloys
Tested at 2000°F and 15,000 psi

Alloy	Life (hours)	Log 10 (Life)	Elongation (%)	R.A. (%)
IV A	25.37	1.4048	2.5	1.6
	32.44	1.5105	2.1	0.8
	24.17	1.3838	2.2	2.0
IV B	3.4	0.5315	4.5	7.3
	2.89	0.4624	6.6	7.7
	2.7	0.4314	4.5	5.8
IV C	10.9	1.0374	4.8	7.7
	14.3	1.1553	7.4	10.8
	10.2	1.0086	5.0	9.3
IV D	6.3*	0.7993	2.6*	4.5
	23.8	1.3766	4.0	5.5
	28.31	1.4518	5.3	6.3
IV E	6.5	0.8129	3.7	7.4
	6.7	0.8261	2.5	4.7
	9.9	0.9956	7.9	12.7
IV F	10.9	1.0374	6.3	9.6
	12.7	1.1038	3.3	5.5
	9.0	0.9542	5.6	5.8
IV G	10.0	1.0000	2.7	2.4
	9.1	0.9590	3.5	2.4
	9.1	0.9590	2.3	2.7
IV H	20.8	1.3181	3.9	3.5
	16.6	1.2201	4.5	5.5
	17.6	1.2455	2.6	2.7
IV I	15.5	1.1903	3.8	3.5
	17.0	1.2304	6.2	4.3
	12.5	1.0969	2.5	8.1
IV J	25.4	1.4048	5.5	7.7
	20.1	1.3032	2.2	4.3
	19.64	1.2923	5.7	8.1
IV K	7.5	0.8751	9.1	16.3
	11.9	1.0755	3.9	11.9
	8.2	0.9138	5.7	10.0
IV L	40.3	1.6053	3.8	8.5
	31.7	1.5011	4.3	4.7
	30.1	1.4786	3.8	9.6

* Results not used in statistical analysis because of the unusual fracture appearance.

TABLE 34

Room Temperature Tensile Results for Series IV Alloys

Alloy	Ultimate (10^3 psi)	0.2% Offset Yield (10^3 psi)	Elongation (%)	R.A. (%)
IV A	124.3	114.0	5.0	6.2
	139.0	125.3	3.0	5.4
IV B	103.5	none	1.0	0.0
	107.6	none	0.1	0.0
IV C	141.2	none	0.8	0.8
	142.3	none	0.4	0.8
IV D	146.4	132.7	2.9	6.3
	137.5	127.1	3.1	5.5
IV E*	112.2	none	1.0	0.0
IV F	133.3	128.9	2.0	7.7
	138.1	-	1.1	4.7
IV G	148.2	144.6	1.1	0.8
	145.8	143.8	1.3	1.2
IV H	136.7	123.3	3.3	8.5
	131.9	117.2	4.6	10.0
IV I	138.1	136.1	2.2	0.8
	137.1	-	1.3	0.8
IV J	135.7	128.1	3.2	9.3
	139.0	128.9	1.9	5.5
IV K	133.7	126.3	2.2	3.9
	127.1	123.5	0.6	7.0
IV L	129.3	116.6	4.3	9.3
	132.4	119.4	4.6	11.2
IV M	139.2	127.7	3.2	3.9
	145.5	138.5	0.9	5.4
IV N	130.7	122.3	2.5	4.7
	131.7	122.7	4.9	9.3
IV O	141.3	129.1	3.4	7.8
	136.5	128.4	2.2	4.3
IV P	132.3	-	0.0	1.6
	133.1	-	0.6	0.8
IV Y	133.9	118.9	4.6	10.1
	132.7	114.4	4.8	9.2
IV Z	134.6	126.4	3.5	5.1
	145.6	141.8	0.9	5.1

* Only one result obtained for this alloy.

TABLE 35

1400°F Tensile Results for Series IV Alloys

Alloy	Ultimate (10^3 psi)	0.2% Offset Yield (10^3 psi)	Elongation (%)	R.A. (%)
IV A	146.8	126.0	4.6	8.1
	149.8	127.7	3.0	6.3
IV B	89.0	-	0.6	0.4
	99.5	-	0.9	0.4
IV C	153.8	147.4	1.9	0.8
	153.6	146.8	1.8	0.8
IV D	149.4	138.0	1.8	2.3
	157.1	135.9	3.9	6.6
IV E*	153.4	-	0.5	1.2
IV F	154.8	143.6	1.8	4.7
	152.0	141.4	2.4	4.7
IV G	161.2	152.0	2.2	2.7
	154.0	147.2	1.4	3.2
IV H	149.0	128.2	2.2	6.6
	153.4	135.8	3.2	7.4
IV I	96.5**	-	1.8**	1.6
	153.2	145.4	2.5	1.2
IV J	147.0	131.0	3.9	9.6
	146.2	129.5	2.9	5.5
IV K	151.8	129.4	2.1	2.7
	147.8	128.3	3.1	4.3
IV L	146.8	122.4	4.3	5.1
	147.8	124.3	2.8	5.5
IV M	155.9	137.5	2.2	4.0
	150.5	138.1	1.4	2.4
IV N	154.8	131.7	6.8	8.5
	149.2	128.2	6.5	10.4
IV O	149.0	129.8	1.5	4.3
	150.4	129.7	1.8	4.7
IV P	126.8	-	1.7	1.2
	123.3	-	2.0	0.8
IV Y	151.4	127.5	3.4	6.2
	153.2	129.8	7.8	14.1
IV Z	152.2	137.7	2.0	2.4
	151.2	136.5	2.5	3.1

* Only one result obtained for this alloy.
** Results not used in statistical analysis because of fracture appearance.

TABLE 36

1875°F Tensile Results for Series IV Alloys

Alloy	Ultimate (10^3 psi)	0.2% Offset Yield (10^3 psi)	Elongation (%)	R.A. (%)
IV A	65.8	59.1	5.4	7.7
	64.1	58.6	6.1	6.2
IV B	53.5	49.6	2.2	2.3
	54.0	49.7	2.9	2.3
IV C	55.3	50.0	6.6	8.5
	53.6	49.3	8.4	13.0
IV D	66.7	60.6	8.7	7.7
	67.1	60.6	4.7	5.8
IV E*	57.3	54.2	4.8	5.4
IV F	54.3	49.5	6.8	9.6
	58.5	53.0	6.9	8.8
IV G	58.5	53.0	6.9	8.8
	56.8	52.4	5.9	7.0
IV H	64.6	58.9	3.2	5.4
	65.9	61.6	3.5	3.1
IV I	59.5	55.7	4.9	5.0
	62.5	58.7	3.6	4.3
IV J	61.2	56.7	7.5	12.2
	62.3	56.8	5.9	5.4
IV K	53.3	47.8	7.3	8.5
	54.0	47.7	7.0	7.7
IV L	62.3	55.5	7.6	8.1
	64.0	55.5	5.6	7.3
IV M	62.4	58.1	4.5	6.2
	62.9	58.0	3.4	4.3
IV N	68.6	63.6	4.5	5.8
	69.1	62.4	3.3	5.8
IV O	55.7	48.9	8.7	10.4
	56.4	49.0	5.9	10.1
IV P	50.0	44.7	6.6	9.6
	51.1	47.5	3.8	7.0
IV Y	69.4	62.5	3.3	3.2
	68.3	63.0	4.0	5.1
IV Z	58.9	53.2	6.4	8.5
	59.5	54.5	6.6	9.2

* Only one result obtained for this alloy.

TABLE 37

2000°F Tensile Results for Series IV Alloys

Alloy	Ultimate (10^3 psi)	0.2% Offset Yield (10^3 psi)	Elongation (%)	R.A. (%)
IV A	42.9	38.7	7.8	9.2
	45.0	40.2	7.7	9.6
IV B	34.8	32.4	6.7	6.2
	35.1	32.6	7.1	6.2
IV C	36.4	33.8	7.2	7.4
	37.3	34.9	7.7	8.1
IV D	47.5	-	6.8	5.1
	47.7	44.2	6.8	6.2
IV E*	37.2	34.2	2.9	1.6
IV F	37.5	34.8	7.7	8.5
	36.8	33.4	10.9	20.2
IV G	38.1	35.7	7.5	8.1
	39.5	36.8	6.9	11.1
IV H	45.5	39.3**	3.4	4.7
	44.5	41.5	5.7	2.4
IV I	43.8	38.2**	3.4	4.3
	44.1	40.7	5.0	3.9
IV J	43.7	37.8**	6.5	7.7
	42.3	37.7	5.4	6.2
IV K	35.1	32.3	8.3	9.6
	34.9	30.9	9.2	10.7
IV L	42.2	38.8	6.3	5.5
	40.8	36.3	6.7	8.5
IV M	43.6	38.8	3.6	3.1
	42.2	37.2	1.2	1.6
IV N	47.9	43.3	6.4	4.7
	49.9	46.2	7.0	8.1
IV O	36.8	33.9	8.1	7.1
	35.2	31.9	8.0	6.6
IV P	33.8	29.9	7.1	6.2
	33.5	29.9	8.9	9.2
IV Y	47.4	-	3.1	3.1
	49.4	45.5	4.8	3.9
IV Z	41.2	36.5	7.0	8.5
	40.5	37.4	7.0	7.4

* Only one result obtained for this alloy.
** 0.1% offset yield.

TABLE 38

Aim and Actual Chemical Components for Series V Alloys *

Alloy	Components Varied at Two Levels				Fixed Components									
	Cr	Ta	W	Hf	C	Mo	Ti	Al	Co	Re	Zr	B	Cb	Ni
V A	4.0	7.5	6.0	1.0	0.13	2.0	1.0	5.4	7.5	0.5	0.13	0.02	0.5	Bal.
	3.93	7.47	5.92	0.98	0.14	2.04	1.05	6.06	7.60	0.62	0.10	0.016	0.49	-
V B	6.0	6.5	4.0	1.75	0.13	2.0	1.0	5.4	7.5	0.5	0.13	0.02	0.5	Bal.
	5.84	7.15	4.26	1.77	0.11	2.05	1.13	6.68	7.08	0.63	0.11	0.020	0.45	-
V C	4.0	6.5	4.0	1.0	0.13	2.0	1.0	5.4	7.5	0.5	0.13	0.02	0.5	Bal.
	4.05	6.49	4.03	0.92	0.12	2.02	1.00	5.72	7.50	0.61	0.14	0.021	0.34	-
V D	4.0	6.5	6.0	1.75	0.13	2.0	1.0	5.4	7.5	0.5	0.13	0.02	0.5	Bal.
	3.90	7.26	5.67	1.90	0.13	2.05	1.11	5.86	7.46	0.61	0.10	0.017	0.38	-
V E	6.0	7.5	4.0	1.0	0.13	2.0	1.0	5.4	7.5	0.5	0.13	0.02	0.5	Bal.
	5.83	7.33	5.21	0.98	0.12	2.00	1.01	5.83	7.04	0.63	0.10	0.017	0.44	-
V F	4.0	7.5	4.0	1.75	0.13	2.0	1.0	5.4	7.5	0.5	0.13	0.02	0.5	Bal.
	3.91	7.34	4.41	1.82	0.11	2.06	1.11	5.97	7.52	0.61	0.12	0.019	0.64	-
V G	6.0	7.5	6.0	1.75	0.13	2.0	1.0	5.4	7.5	0.5	0.13	0.02	0.5	Bal.
	5.78	7.16	5.97	1.82	0.13	2.06	1.03	5.66	6.98	0.62	0.12	0.020	0.49	-
V H	6.0	6.5	6.0	1.0	0.13	2.0	1.0	5.4	7.5	0.5	0.13	0.02	0.5	Bal.
	5.86	7.10	5.82	1.03	0.13	2.09	1.01	5.87	7.12	0.63	0.12	0.018	0.41	-

* Aim is shown first. All data is in weight percent. Trace Elements: Fe $<$ 0.30, Mn $<$ 0.05, Si $<$ 0.15, and S $<$ 0.01.

TABLE 39

Stress Rupture Data for Series V Alloys
Tested at 2000°F and 15,000 psi

Alloy	Life (hours)	Log 10 (Life)	Elongation (%)	R.A. (%)
V A	18.9	1.2765	8.0	11.2
	25.6	1.4082	10.0	13.1
	25.4	1.4048	12.7	14.9
V B	14.0	1.1461	14.0	16.0
	12.9	1.1106	15.5*	11.5
	16.1	1.2068	12.7	10.4
V C	8.8	0.9445	19.2	20.0
	9.1	0.9590	15.2	15.0
	7.7	0.8865	13.9	18.5
V D	14.6	1.1644	18.8*	13.8
	14.7	1.1673	12.3	13.8
	10.7	1.0294	9.9	13.8
V E	19.3	1.2856	13.2	13.1
	23.0	1.3617	9.7	13.1
	22.6	1.3541	10.0	9.8
V F	1.62**	0.2095	10.2**	10.4
	14.9	1.1732	10.9	16.0
	12.8	1.1072	7.3	10.4
V G	43.0	1.6335	6.8***	4.3
	38.2	1.5821	6.0	4.7
	44.3	1.6464	6.9	12.6
V H	23.5	1.3711	8.5	9.3
	21.55	1.3345	12.0***	7.8
	33.6	1.5263	11.8	9.3

* 3 piece fracture and specimen bent due to cracking.

** Results not used in statistical analysis.

*** 3 piece fracture.

TABLE 39

Stress Rupture Data for Series V Alloys
Tested at 2000°F and 15,000 psi

lloy	Life (hours)	Log 10 (Life)	Elongation (%)	R.A. (%)
A	18.9	1.2765	8.0	11.2
	25.6	1.4082	10.0	13.1
	25.4	1.4048	12.7	14.9
B	14.0	1.1461	14.0	16.0
	12.9	1.1106	15.5*	11.5
	16.1	1.2068	12.7	10.4
C	8.8	0.9445	19.2	20.0
	9.1	0.9590	15.2	15.0
	7.7	0.8865	13.9	18.5
D	14.6	1.1644	18.8*	13.8
	14.7	1.1673	12.3	13.8
	10.7	1.0294	9.9	13.8
E	19.3	1.2856	13.2	13.1
	23.0	1.3617	9.7	13.1
	22.6	1.3541	10.0	9.8
F	1.62**	0.2095	10.2**	10.4
	14.9	1.1732	10.9	16.0
	12.8	1.1072	7.3	10.4
G	43.0	1.6335	6.8***	4.3
	38.2	1.5821	6.0	4.7
	44.3	1.6464	6.9	12.6
H	23.5	1.3711	8.5	9.3
	21.55	1.3345	12.0***	7.8
	33.6	1.5263	11.8	9.3

3 piece fracture and specimen bent due to cracking.

** Results not used in statistical analysis.

*** 3 piece fracture.

TABLE 40

Room Temperature Tensile Results for Series V Alloys

Alloy	Ultimate (10^3 psi)	0.2% Offset Yield (10^3 psi)	Elongation (%)	R.A. (%)
V A	137.8	116.8	5.7	8.1
	141.7	118.8	7.3	9.7
V B	132.4	114.2	5.0	7.4
	141.5	114.5	6.6	6.6
V C	141.7	104.9	7.3	9.3
	148.3	103.9	7.3	11.9
V D	130.8	110.5	5.4	7.4
	129.6	108.9	6.6	10.8
V E	139.2	121.3	4.3	6.2
	141.7	123.1	6.1	10.1
V F	137.5	108.3	6.2	8.5
	135.4	109.4	4.4	9.7
V G	130.7	123.9	1.7	7.0
	141.8	130.1	3.5	7.0
V H	141.1	121.2	7.1	10.7
	145.5	120.7	8.1	8.5

TABLE 41

1400°F Tensile Results for Series V Alloys

Alloy	Ultimate (10^3 psi)	0.2% Offset Yield (10^3 psi)	Elongation (%)	R.A. (%)
V A	134.9	120.4	2.0	4.3
	136.2	120.7	1.8	3.2
V B	136.5	109.6	3.4	4.7
	131.1	108.2	2.6	5.9
V C	136.2	122.9	1.3	5.1
	129.7	122.2	2.1	4.3
V D	125.2	112.0	1.3	4.3
	132.7	118.1	1.9	5.5
V E	142.3	116.1	2.2	4.7
	136.9	118.3	2.2	5.5
V F	144.4	118.9	1.7	5.1
	145.2	125.4	2.8	3.5
V G	154.0	126.8	4.1	5.9
	153.0	125.5	3.7	8.6
V H	140.1	113.3	3.4	6.3
	142.8	114.4	3.6	7.4

TABLE 42

1875°F Tensile Results for Series V Alloys

Alloy	Ultimate (10^3 psi)	0.2% Offset Yield (10^3 psi)	Elongation (%)	R.A. (%)
V A	57.3	47.6	9.7	10.9
	58.4	46.4	8.4	8.6
V B	53.8	44.4	19.2	20.6
	53.9	43.4	14.7	16.3
V C	46.3	38.6	16.4	16.4
	48.1	39.1	13.8	13.9
V D	54.6	44.6	9.4	8.9
	54.8	46.1	10.0	7.8
V E	56.0	44.8	12.7	12.3
	56.0	45.3	10.3	10.4
V F	54.4	45.8	9.0	9.6
	56.4	46.9	7.7	8.5
V G	61.8	53.8	7.7	7.8
	64.5	55.2	7.0	6.6
V H	54.5	45.4	12.3	10.8
	55.3	45.5	10.0	13.1

TABLE 43

2000°F Tensile Results for Series V Alloys

Alloy	Ultimate (10^3 psi)	0.2% Offset Yield (10^3 psi)	Elongation (%)	R.A. (%)
V A	36.7	30.6	11.8	7.8
	36.9	32.0	12.4	10.5
V B	33.8	28.2	17.3	22.3
	33.5	28.6	16.4	13.0
V C	28.5	24.9	21.0	20.3
	28.6	25.1	23.8	18.7
V D	34.4	28.6	10.9	10.0
	34.6	29.1	11.1	10.0
V E	35.7	30.4	12.8	11.5
	35.8	30.2	10.8	10.1
V F	35.6	30.0	8.2	6.6
	35.0	29.8	10.8	9.6
V G	41.8	37.7	6.7	6.2
	41.0	36.9	7.3	9.6
V H	34.8	30.4	12.9	10.1
	34.3	31.1	12.5	12.0

TABLE 44

Aim and Actual Chemical Compositions for Series VI Alloys *

Alloy	Co	Cr	Zr	B	C	Ni	Mo	W	Ta	Ti	Al	Re	Cb	V	Hf
VI A	7.5 6.95	6.1 6.21	0.13 0.13	0.02 0.020	0.13 0.11	Bal. -	2.0 2.09	5.8 6.32	9.0 8.26	1.0 0.87	5.4 5.27	0.5 0.55	0.5 0.45	0 -	0.43 0.46
VI B	5.0 4.78	6.0 5.80	0.08 0.068	0.02 0.022	0.15 0.14	Bal. -	3.0 2.61	6.0 6.38	8.5 8.51	1.0 0.97	5.4 5.35	0.5 0.60	1.0 0.94	0.5 0.51	1.75 1.78
VI C	5.0 5.03	6.0 6.16	0.08 0.073	0.02 0.021	0.15 0.15	Bal. -	2.0 2.08	6.0 6.43	8.0 8.55	1.3 1.31	5.4 5.49	0.5 0.54	0.75 0.66	0 -	1.75 1.84
VI D	5.0 4.95	5.4 5.77	0.08 0.078	0.02 0.019	0.15 0.16	Bal. -	2.0 2.01	6.0 5.81	8.5 8.01	1.0 1.03	5.4 5.18	0 -	0.5 0.51	0.5 0.55	1.75 1.72

* Aim is shown first. All data is in weight percent. Trace Elements: Fe $<$ 0.30, Mn $<$ 0.05, Si $<$ 0.15, and S $<$ 0.01.

TABLE 45

Stress Rupture Data for Series VI Alloys
Tested at 2000°F and 15,000 psi

Alloy	Life (hours)	Log 10 (Life)	Elongation (%)	R.A. (%)
VI A	61.8	1.7910	6.4	8.5
	59.9	1.7774	7.7	8.5
	66.9	1.8254	9.3	9.3
VI B	39.6	1.5977	4.4	5.1
	35.36	1.5485	4.9	5.9
	42.05	1.6237	4.2	7.4
VI C	33.95	1.5308	3.2	3.9
	37.0	1.5682	3.9	7.4
	35.93	1.5555	4.5	5.1
VI D	68.0	1.8325	3.0	8.5
	53.54*	1.7287	3.3	2.4
	80.3	1.9047	7.2**	7.4

* Temperature varied 10-30° higher than test temperature on this specimen.

** 3 piece fracture.

TABLE 46

Room Temperature Tensile Results for Series VI Alloys

Alloy	Ultimate (10^3 psi)	0.2% Offset Yield (10^3 psi)	Elongation (%)	R.A. (%)
VI A	139.4	130.1	2.6	3.9
	156.7	137.4	4.9	5.9
VI B	141.0	131.7	2.6	6.2
	137.9	128.5	2.6	5.5
VI C	146.4	133.6	1.9	5.1
	144.6	129.1	3.7	5.1
VI D	142.4	131.8	3.1	6.6
	141.8	131.6	2.1	4.7

TABLE 47

1400°F Tensile Results for Series VI Alloys

Alloy	Ultimate (10^3 psi)	0.2% Offset Yield (10^3 psi)	Elongation (%)	R.A. (%)
VI A	158.2	135.8	4.7	5.1
	157.6	136.0	5.0	7.0
VI B	158.8	140.6	3.8	5.5
	155.8	139.9	2.6	5.5
VI C	158.6	142.0	2.5	3.9
	159.1	141.5	3.6	7.7
VI D	155.2	136.1	4.9	7.7
	134.8	131.3	2.2	3.1

TABLE 48

1875°F Tensile Results for Series VI Alloys

Alloy	Ultimate (10^3 psi)	0.2% Offset Yield (10^3 psi)	Elongation (%)	R.A. (%)
VI A	68.8	58.9	6.8	8.5
	70.0	58.4	6.9	7.8
VI B	69.8	62.4	8.2	15.3
	69.0	63.5	6.1	10.1
VI C	69.4	63.4	6.0	8.1
	69.3	63.1	4.9	6.6
VI D	70.5	62.6	6.6	7.0
	70.3	63.2	6.1	8.5

TABLE 49

2000°F Tensile Results for Series VI Alloys

Alloy	Ultimate (10^3 psi)	0.2% Offset Yield (10^3 psi)	Elongation (%)	R.A. (%)
VI A	48.6	48.6	5.9	5.9
	48.6	41.9	5.8	5.5
VI B	48.9	44.7	5.7	6.2
	47.7	44.1	4.9	8.1
VI C	50.7	46.7	1.8	2.3
	49.7	46.9	4.3	3.5
VI D	48.5	44.7	7.6	7.0
	50.1	45.6	7.9	8.1

TABIE 50

Workability Studies for Series I

Alloy	Rating	Average 2000°F S. R. Life(hrs.)	Number of Pieces Pressed	Sidepressing Procedure 2050°F 50% Red.	2050°F 25% Red.	2100°F 50% Red.	2150°F 50% Red.
I 1	2.5	None	2	x		x	
I 2	2.0	0.07	3	x		x	x
I 3	2.0	0.11	3	x		x	x
I 4	3.0	0.25	2	x		x	
I 5	3.5	11.9	2	x		x	
I 6	3.5	8.1	2	x		x	
I 7	4.0	16.4	1		x		
I 8	4.0	4.6	2	x		x	
I 9	4.0	1.7	2	x		x	
I 10	2.5	0.24	4	x	x	x	x
I 11	3.0	4.1	4	x	x	x	x
I 12	2.5	2.4	3	x	x	x	
I 13	3.0	8.5	4	x	x	x	x
I 14	4.0	4.1	3	x	x	x	
I 15	5.0	2.5	2	x		x	
I 16	3.0	1.0	3	x	x	x	
I 17	3.5	2.4	4	x	x	x	x
I 18	4.0	6.0	3	x	x	x	
I 19	3.0	5.7	4	x	x	x	x
I 20	4.0	2.5	4	x	x	x	x
I 21	3.5	2.3	2	x		x	
I 22	3.0	0.26	4	x	x	x	x
I 23	3.0	1.3	4	x	x	x	x
I 24	3.5	2.6	3	x	x	x	
I 25	3.5	1.6	3	x	x	x	
I 26	4.0	0.37	4	x	x	x	x
I 27	5.0	0.23	4	x	x	x	x

TABLE 51

Workability Studies for Series II

Alloy	Rating	Average 2000°F S. R. Life (hrs.)	Number of Pieces Pressed	Sidepressing Procedure 2100°F 30% Red.	Sidepressing Procedure 2100°F 20% Red.	Sidepressing Procedure 2150°F 30% Red.
IIa	2.5	7.9	2	x		x
IIb	3.0	14.3	2	x		x
IIc	3.5	13.2	2	x		x
IId	3.5	17.1	2	x		x
IIe	4.0	13.7	1	x		
IIf	4.5	5.0	2	x		x
IIg	5.0	19.2	3	x	x	x
IIh	3.5	8.8	2	x		x
IIi	5.0	7.0	2	x		x

TABLE 52

Workability Studies for Series III

				Sidepressing Procedure			
				2100°F		2150°F	
Alloy	Rating	Average 2000°F S. R. Life (hrs.)	Number of Pieces Pressed	30% Red.	20% Red.	30% Red.	20% Red.
IIIa	3.5	7.3	2	x		x	
IIIb	3.0	8.7	3	x	x	x	
IIIc	3.0	8.6	2	x		x	
IIId	3.5	16.6	3	x	x	x	
IIIe	3.0	11.3	3	x	x	x	
IIIf	4.0	5.0	3	x	x	x	
IIIg	3.0	15.5	2	x		x	
IIIh	5.0	4.2	3	x		x	x
IIIi	6.0	11.4	0				

TABLE 53

Workability Studies for Series IV

Alloy	Rating	Average 2000°F S. R. Life (hrs.)	Number of Pieces Pressed	Sidepressing Procedure 2100°F 15% Red.	25% Red.	30% Red.	35% Red.	2150°F 30% Red.
IV A	6.0	27.3	0					
IV B	5.0	3.0	3	x		x		x
IV C	4.0	11.8	3		x		x	x
IV D	4.5	26.1	2			x		x
IV E	4.0	7.7	2			x		x
IV F	3.5	10.9	3			xx		x
IV G	4.0	9.4	1			x		
IV H	4.0	18.3	1			x		
IV I	5.0	15.0	2			x		x
IV J	6.0	21.7	0					
IV K	4.0	9.2	2			x		x
IV L	3.5	34.0	1			x		
IV M	6.0	23.7	0					
IV N	6.0	33.2	0					
IV O	3.5	16.0	1			x		
IV P	4.0	5.6	2			x		x
IV Y	4.0	41.3	1			x		
IV Z	4.5	15.1	2			x		x

TABLE 54

Workability Studies for Series V

Alloy	Rating	Average 2000°F S. R. Life (hrs.)	Number of Pieces Pressed	Sidepressing Procedure 2100°F 15% Red.	Sidepressing Procedure 2100°F 30% Red.	Sidepressing Procedure 2150°F 15% Red.	Sidepressing Procedure 2150°F 30% Red.
V A	6.0	23.3	0				
V B	4.0	14.3	2		x		x
V C	4.5	8.5	5	x	x	xx	x
V D	4.0	13.3	1		x		
V E	6.0	21.6	0				
V F	4.0	13.8	1		x		
V G	6.0	41.8	0				
V H	4.0	26.2	2		x		x

TABLE 55

Workability Studies for Series VI

Alloy	Rating	Average 2000°F S. R. Life (hrs.)	Number of Pieces Pressed
VI A	6.0	62.9	0
VI B	6.0	39.0	0
VI C	6.0	35.6	0
VI D	6.0	67.3	0

TABLE 56 (Continued)

Main Effects and Interactions on Mechanical Properties Produced by Increasing Concentration of an Element for Series IV Alloys

Estimated Effects *	2000°F Tensile (psi)	2000°F Elong. (%)	2000°F Stress Rupture Life (Log_{10} (Life in hrs.))	2000°F Stress Rupture Elong. (%)
Co	-1012	1.262	-0.0902	0.554
Cr	-8400	1.712	-0.4634	1.371
CoCr+TaW-HfC-ReZr	-1588	-0.050	-0.0536	0.338
Ta	212	-2.188	-0.2228	-0.679
CoTa+CrW-ReC-HfZr	600	1.125	-0.0349	-0.196
CrTa+CoW+HfRe+CZr	-312	0.050	-0.0744	-1.412
W	-1700	0.062	-0.1382	-0.262
Hf	488	-0.612	-0.1382	0.304
CoHf+WRe-CrC-TaZr	-1300	-0.500	-0.0538	0.238
CrHf-CoC+TaRe-WZr	-738	-0.100	-0.0229	0.238
C	-1700	-1.012	0.0425	0.296
TaHf-CoZr+CrRe-WC	-950	-0.850	-0.0399	0.554
Zr	-1112	-0.288	-0.0217	0.596
Re	1050	-1.138	0.0204	0.838
CoRe+HfW-TaC-CrZr	-62	0.800	-0.0116	0.721
Standard Error of Each Effect	290	0.345	0.0271	0.618
99% Significance Level of Each Effect ≥	847 **	1.007 **	0.0745 ***	1.700 ***

* Positive sign refers to an increase in weight percent of the element.
** 16 degrees of freedom
*** 32 degrees of freedom

TABLE 56

Main Effects and Interactions on Mechanical Properties Produced by Increasing Concentration of an Element for Series IV Alloys

Estimated Effects*	Room Temp. Tensile(psi)	Room Temp. Elong.(%)	1400°F Tensile(psi)	1400°F Elong.(%)	1875°F Tensile(psi)	1875°F Elong(%)
Co	-323	0.281	-1321	-0.144	-1075	0.056
Cr	-538	-2.006	-852	-1.769	-9338	0.819
CoCr+TaW-HfC-ReZr	-404	-0.106	-1162	0.294	-1875	-0.894
Ta	-301	-0.906	-746	-1.356	-525	-1.819
CoTa+CrW-ReC-HfZr	-293	0.169	-1021	0.531	538	-0.456
CrTa+CoW+HfRe+CZr	-930	0.081	-1162	0.531	200	-0.644
W	-26	-1.019	-1082	-0.531	-2912	-0.431
Hf	-1035	-0.244	-428	0.256	200	-1.231
CoHf+WRe-CrC-TaZr	95	-0.169	-433	-0.431	-1212	-0.069
CrHf-CoC+TaRe-WZr	-876	0.144	-452	-0.556	400	-0.031
C	-34	0.306	222	-0.656	-1012	0.069
TaHf-CoZr+CrRe-WC	-823	0.419	498	-0.719	-312	-0.0594
Zr	-23	-0.594	283	-0.269	-2100	1.106
Re	-361	-0.294	-372	-0.181	1138	-0.269
CoRe+HfW-TaC-CrZr	229	0.106	-392	-0.394	-125	0.719
Standard Error of Each Effect	139	0.316	117	0.236	416	0.433
99% Significance ≙ Level of Each Effect	406**	0.923 **	342 **	0.689 **	1215 **	1.264 **

* Positive sign refers to an increase in weight percent of the element

** 16 degrees of freedom

TABLE 57

Main Effects and Interactions on Mechanical Properties Produced by Increasing Concentration of an Element for Series V Alloys

Estimated Effects *		Room Temp. Tensile(psi)	Room Temp. Elong.(%)	1400°F Tensile(psi)	1400°F Elong.(%)	1875°F Tensile(psi)	1875°F Elong.(%)
Cr		139	-0.975	652	1.288	3188	1.175
W		-234	-0.255	208	0.438	4538	-3.650
Cr(-W)+Ta(-Hf)		-341	0.175	-870	-0.662	438	1.325
Ta		-64	-1.775	908	0.112	5438	-4.150
CrTa+(-W)(-Hf)		-114	-1.025	-15	-0.312	-238	-0.475
(-W)Ta+Cr(-Hf)		-189	0.475	-25	-0.238	-262	-1.950
Hf		-716	-1.725	288	0.362	2788	-1.125
Standard Error of Each Effect		215	0.520	162	0.220	507	0.815
99% Significance Level of Each Effect	≥	722**	1.747**	544**	0.739**	1704**	2.738**

Estimated Effects *		2000°F Tensile(psi)	2000°F Elong.(%)	2000°F Stress Rupture Life (Log_{10}(Life in hrs.))	2000°F Stress Rupture Elong.(%)
Cr		2550	-1.662	0.2415	-1.683
W		3500	-4.438	0.2391	-2.250
Cr(-W)+Ta(-Hf)		225	0.0375	-0.0324	1.600
Ta		4500	-5.638	0.2106	-4.433
CrTa+(-W)(-Hf)		-25	0.262	-0.0159	0.783
(-W)Ta+Cr(-Hf)		-75	-3.338	-0.0158	-0.617
Hf		2300	-3.662	-0.0005	-1.167
Standard Error of Each Effect		150	0.564	0.0354	1.406
99% Significance Level of Each Effect	≥	504**	1.895**	0.1034***	4.106***

* Positive sign refers to an increase in weight percent of the element
** 8 degrees of freedom
*** 16 degrees of freedom

TABLE 58

Rockwell C Hardness Results for Several Alloys After Tensile Tests at Room Temperature and 1400°F

Alloy	Condition	Hardness, R_c
IV L	R. T.	42.6
	1400°F	42.3
IV N	R. T.	42.4
	1400°F	45.6
IV Y	R. T.	46.3
	1400°F	45.9
V G	R. T.	43.0
	1400°F	43.5

TABLE 59

Main Effects and Interactions on Workability Results Produced by Increasing Concentration of an Element for Series IV Alloys

Estimated Effects *	Ratings **
Co	0.125
Cr	-1.125
CoCr+TaW-HfC-ReZr	0.125
Ta	0
CoTa+CrW-ReC-HfZr	-0.500
CrTa+CoW+HfRe+CZr	0.500
W	0.750
Hf	0.250
CoHf+WRe-CrC-TaZr	0
CrHf-CoC+TaRe-WZr	0
C	-0.500
TaHf-CoZr+CrRe-WC	-0.375
Zr	-0.375
Re	0.625
CoRe+HfW-TaC-CrZr	-0.375

* Positive sign refers to an increase in weight percent of the element.

** Positive rating means that the effect is detrimental to workability.

TABLE 60

Main Effects and Interactions on Workability Results Produced by Increasing Concentration of an Element for Series V Alloys

Estimated Effects *	Ratings **
Cr	0.375
W	0.375
Cr(-W)+Ta(-Hf)	0.375
Ta	1.375
CrTa+(-W)(-Hf)	0.625
(-W) Ta+Cr(-Hf)	-0.625
Hf	-0.625

* Positive sign refers to an increase in weight percent of the element.

** Positive rating means that the effect is detrimental to workability.

TABLE 61

Pertinent Melting Data for Task II Alloys

Alloy*	Virgin Heats: Total Time, hrs.	Virgin Heats: "Silverfish" Temperature, °F	Remelt Heats: Total Time, hrs.	Remelt Heats: Pressure at 2900°F, μ
Cast				
IV Y-1	1:15	2490	0:56	10
-2	1:37	2475	0:52	3
-3	1:17	2480	0:43	8
VI A-1	2:08	2455	0:50	8
-2	1:25	2490	0:54	10
-3	1:40	2435	0:44	4
VI D-1	1:26	2485	1:01	10
-2	1:56	2460	1:17	10
-3	2:18	2405	0:40	4
Wrought				
I-5	1:30	2470	0:42	N.A.**
II b	1:17	2420	0:55	N.A.
II d	1:31	2435	0:42	N.A.
III g	1:22	2430	0:49	9
IV Y	1:12	2455	1:44	8

* Three 50 lb. heats were made for each cast alloy, and one 50 lb. heat for each wrought alloy.

** Not available.

TABLE 62

Aim and Actual Chemical Compositions for Task II Cast Alloys*

Alloy	C	Cr	Mo	Ti	Al	Co	W	Re	Hf	Zr	B	Ta	Cb	V	Ni
IV Y (Aim)	0.15	6.0	2.0	1.0	5.4	5.0	5.5	1.0	2.0	0.03	0.02	8.0	1.0	0	Bal.
1	0.16	6.03	2.13	1.03	5.29	4.93	5.52	1.04	2.05	0.040	0.024	8.35	0.96	-	-
2	0.11	5.88	2.02	1.01	5.22	4.78	5.47	0.90	2.01	0.043	0.027	8.22	0.92	-	-
3	0.15	5.85	2.04	1.08	5.25	4.82	5.36	0.90	2.01	0.039	0.023	8.17	0.89	-	-
VI A (Aim)	0.13	6.1	2.0	1.0	5.4	7.5	5.8	0.5	0.43	0.13	0.02	9.0	0.5	0	Bal.
1	0.14	6.04	2.23	0.87	5.26	6.86	5.76	0.55	0.52	0.12	0.022	8.49	0.33	-	-
2	0.16	6.03	2.24	0.98	5.26	7.19	5.78	0.57	0.46	0.12	0.018	8.32	0.34	-	-
3	0.11	6.06	2.20	0.97	5.07	7.16	5.80	0.56	0.55	0.12	0.025	8.52	0.34	-	-
VI D (Aim)	0.15	5.4	2.0	1.0	5.4	5.0	6.0	0	1.75	0.08	0.02	8.5	0.5	0.5	Bal.
1	0.14	5.80	2.16	1.03	5.12	4.84	5.67	-	1.96	0.11	0.024	8.08	0.33	0.45	-
2	0.11	5.83	2.18	1.05	5.10	4.84	5.85	-	2.07	0.10	0.023	8.10	0.34	0.45	-
3	0.12	5.91	2.21	1.04	5.25	4.86	5.66	-	2.03	0.11	0.033	7.89	0.34	0.50	-

* Three 50 lb. heats were made for each alloy. Aim is shown first and then the actual chemical compositions for each of the three heats is given. All data is in weight percent. Trace Elements: Fe $<$ 0.30, Mn $<$ 0.01, Si $<$ 0.03, and S $<$ 0.01.

TABLE 63

Aim and Actual Chemical Compositions for Task II Wrought Alloys*

Alloy	C	Cr	Mo	Ti	Al	Co	W	Re	Hf	Zr	B	Ta	Cb	V	Ni
I-5 (Aim)	0.13	10.0	1.0	1.0	6.3	10.0	5.5	0	0	0.03	0.02	4.5	0	0	Bal.
	0.16	11.19	1.30	0.95	5.96	9.87	5.13	-	-	0.018	0.025	4.60	-	-	-
IIb (Aim)	0.13	10.0	2.0	1.0	4.5	10.0	5.5	0	1.0	0.03	0.02	8.0	0	1.0	Bal.
	0.12	9.73	2.26	1.12	4.47	9.45	5.62	-	1.10	0.033	0.026	8.27	-	0.70	-
IId (Aim)	0.13	10.0	2.0	1.0	4.5	10.0	5.5	0	1.0	0.03	0.02	8.0	1.0	0	Bal.
	0.12	9.91	1.96	0.94	4.55	9.37	5.54	-	0.88	0.024	0.019	7.91	0.68	-	-
IIIg (Aim)	0.13	10.0	2.0	1.0	4.5	10.0	5.5	4.0	0	0.03	0.02	8.0	0	0	Bal.
	0.15	8.95	2.01	0.81	4.23	9.32	5.66	2.90	-	0.020	0.028	7.58	-	-	-
IV Y (Aim)	0.15	6.0	2.0	1.0	5.4	5.0	5.5	1.0	2.0	0.03	0.02	8.0	1.0	0	Bal.
	0.10	5.91	2.04	1.01	5.26	4.85	5.54	0.98	2.08	0.038	0.022	8.40	0.97	-	-

* All data are in weight percent.
Trace Elements: Fe < 0.30, Mn < 0.01, Si < 0.03, and S < 0.01.

TABLE 64

Stress Rupture Results for Task II Cast Alloys Tested in Air Atm. at 1400°F

	85,000 psi			90,000 psi			94,000 psi		
Alloy	Life (hours)	Elongation (%)	R.A. (%)	Life (hours)	Elongation (%)	R.A. (%)	Life (hours)	Elongation (%)	R.A. (%)
IV Y	355.3	1.6	2.0	146.4	0.6	0.8	0.2	0	0
	777.7	3.7	12.8	617.1	2.5	3.1	620.0	1.2	5.2
	1309.6	2.7	7.4	371.8	3.5	3.4	406.2	3.0	3.9
VI A	592.6	3.1	3.2	72.6	0.4	2.0	144.9	1.8	3.9
	943.6	1.7	3.1	641.1	2.0	2.8	400.4	2.0	2.4
	1355.5	1.9	2.8	665.4	1.3	2.2	516.7	2.9	7.2
VI D	534.9	1.0	2.4	49.5	1.7	3.7	214.2	1.9	4.0
	B.O.L.*	-	-	272.6	1.5	2.3	19.7	0.8	2.0
	1124.1	3.1	2.0	546.7	2.2	5.5	312.4	2.0	6.1

* Broke on Loading.

TABLE 65

Stress Rupture Results for Task II Cast Alloys Tested in Air Atm. at 1875°F

	15,000 psi			25,000 psi			35,000 psi		
Alloy	Life (hours)	Elongation (%)	R.A. (%)	Life (hours)	Elongation (%)	R.A. (%)	Life (hours)	Elongation (%)	R.A. (%)
IV Y	468.2	5.0	5.5	29.9	4.6	3.1	6.3	3.4	0.4
	842.9	3.3	3.9	76.9	4.1	5.6	13.3	7.7	2.8
	252.2	6.5	14.4	78.3	8.6	4.7	11.7	6.9	8.6
VI A	1174.3	4.0	8.2	69.1	5.9	7.0	9.4	2.4	3.9
	693.4	5.7	9.7	54.4	5.1	3.9	8.7	4.8	2.4
	771.3	8.0	10.0	82.7	8.3	14.4	13.3	4.5	7.0
VI D	721.3	8.5	8.5	10.3	4.1	6.3	6.2	2.5	1.2
	799.0	4.6	4.1	31.9	2.4	3.1	3.8	3.1	0.8
	667.5	4.3	5.6	57.6	6.1	6.6	8.1	4.2	7.8

TABLE 66

Stress Rupture Results for Task II Cast Alloys Tested in Air Atm. at 1950°F

	15,000 psi			20,000 psi			25,000 psi		
Alloy	Life (hours)	Elongation (%)	R.A. (%)	Life (hours)	Elongation (%)	R.A. (%)	Life (hours)	Elongation (%)	R.A. (%)
IV Y	77.7	2.5	1.0	8.4	2.7	1.6	15.7	3.0	2.5
	217.8	0.9	2.7	39.2	5.5	8.0	14.6	3.5	8.2
	101.1	4.3	6.6	2.1	2.3	2.4	6.2	2.7	6.3
VI A	187.9	6.2	5.1	40.6	4.5	7.0	1.2	3.8	5.1
	156.1	5.1	7.4	33.1	6.5	8.0	12.1	5.9	7.4
	208.2	7.5	13.0	55.1	10.3	11.5	83.8	8.6	8.5
VI D	114.3	7.6	11.5	29.6	4.4	5.4	14.6	4.1	9.0
	149.2	2.7	3.9	24.5	1.8	3.1	6.3	2.1	3.1
	151.8	4.5	6.0	28.6	5.1	7.8	0.8	3.1	5.3

TABLE 67

Stress Rupture Results for Task II Cast Alloys Tested in Argon Atm. Under a 15,000 psi Load

	1875°F			1950°F		
Alloy	Life (hours)	Elongation (%)	R.A. (%)	Life (hours)	Elongation (%)	R.A. (%)
IV Y	298.1	4.2	7.7	66.0	1.7	2.0
	967.7	4.1	6.4	177.9	6.4	3.3
	683.7	6.1	6.3	97.2	4.4	6.6
VI A	704.0	5.1	8.5	195.7	6.5	7.6
	650.0	4.4	6.3	140.9	8.0	8.9
	788.5	7.5	18.1	164.9	7.2	4.1
VI D	391.8	2.3	6.8	145.5	7.3	9.3
	0.9	0.6	1.6	109.6	4.1	2.0
	739.5	9.4	14.5	145.0	6.7	10.0

TABLE 68

Tensile Results for Task II Cast Alloys

Alloy	Room Temperature				1200°F			
	Ultimate (10^3 psi)	0.2% Offset Yield(10^3 psi)	Elongation (%)	R. A. (%)	Ultimate (10^3 psi)	0.2% Offset Yield(10^3 psi)	Elongation (%)	R. A. (%)
IV Y	146.0	136.6	1.9	7.4	157.4	132.3	7.0	12.3
	140.9	134.0	3.2	7.4	163.0	134.0	4.7	8.1
VI A	154.6	135.4	4.7	7.4	168.8	137.4	4.8	7.4
	157.0	139.5	4.5	7.0	163.8	137.2	3.0	6.6
VI D	148.2	132.0	2.6	5.5	153.8	136.0	2.9	5.1
	150.6	131.3	3.8	5.5	155.0*	138.4	1.4	1.2

* Broke in shoulder

TABLE 68 (Continued)

Tensile Results for Task II Cast Alloys

	1400°F				1600°F			
Alloy	Ultimate (10^3 psi)	0.2% Offset Yield(10^3psi)	Elongation (%)	R. A. (%)	Ultimate (10^3 psi)	0.2% Offset Yield(10^3psi)	Elongation (%)	R. A. (%)
IV Y	155.4	134.8	4.9	8.1	124.2	115.4	7.6	5.9
	163.0	139.1	4.4	6.6	118.3	115.1	4.5	9.6
VI A	159.8	138.0	3.8	5.5	126.5	-	2.8	5.1
	161.4	139.2	4.0	4.3	126.4	112.0	2.2	3.1
VI D	156.8	140.4	1.3	4.7	126.0	115.0	2.2	4.3
	157.4	137.0	4.3	6.6	125.6	115.6	3.4	3.9

TABLE 68 (Continued)

Tensile Results for Task II Cast Alloys

	1875°F				1950°F			
Alloy	Ultimate (10^3 psi)	0.2% Offset Yield(10^3psi)	Elongation (%)	R. A. (%)	Ultimate (10^3 psi)	0.2% Offset Yield(10^3psi)	Elongation (%)	R. A. (%)
IV Y	69.8	61.0	6.5	6.6	58.2	52.0	6.3	8.1
	71.1	61.8	5.9	5.9	58.0	51.9	5.2	6.2
VI A	73.0	62.1	4.4	5.5	57.1	50.0	6.0	5.9
	73.0	63.2	3.9	4.7	57.4	50.4	3.8	2.7
VI D	73.6	65.5	3.3	9.6	60.0	54.8	2.1	2.0
	74.1	66.6	2.2	5.9	61.2	56.1	2.3	2.4

TABLE 69

Tensile Results for Task II Cast Alloys after a 1875°F/300 hour Argon Age

	Room Temperature				1200°F			
Alloy	Ultimate (10^3 psi)	0.2% Offset Yield(10^3psi)	Elongation (%)	R.A. (%)	Ultimate (10^3 psi)	0.2% Offset Yield(10^3psi)	Elongation (%)	R.A. (%)
IV Y	122.9	114.6	2.2	5.5	136.9	116.3	2.7	4.7
	128.8	115.3	1.0	1.2	117.3	107.0	2.3	4.3
VI A	149.6	120.4	4.2	7.4	138.1	113.8	3.4	7.7
	112.0	111.2	1.4	0.8	128.7	110.6	4.0	7.0
VI D	117.4	105.4	2.9	8.5	141.1	118.4	2.6	7.0
	112.6	108.2	1.9	3.9	121.5	-	1.6	4.0

TABLE 69 (Continued)

Tensile Results for Task II Cast Alloys after a 1875°F/300 hours Argon Age

	1400°F				1600°F			
Alloy	Ultimate (10^3 psi)	0.2% Offset Yield (10^3 psi)	Elongation (%)	R.A. (%)	Ultimate (10^3 psi)	0.2% Offset Yield (10^3 psi)	Elongation (%)	R.A. (%)
IV Y	121.3	113.7	2.1	4.3	103.5	-	0.6	2.0
	150.0	139.6	1.1	1.6	113.0	110.6	1.7	1.2
VI A	117.2	116.2	1.0	3.1	101.7	-	1.4	1.2
	128.6	113.4	3.4	5.6	117.5	115.1	2.6	2.0
VI D	128.9	121.3	1.0	4.3	92.9	-	1.1	1.2
	124.1	121.1	1.3	3.1	113.6	110.2	1.5	3.5

TABLE 69 (Continued)

Tensile Results for Task II Cast Alloys after a 1875°F/300 hours Argon Age

Alloy	1875°F Ultimate (10^3psi)	0.2% Offset Yield (10^3psi)	Elongation (%)	R.A. (%)
IV Y	60.3	51.2	4.8	4.3
	64.6	60.4	1.6	3.5
VI A	52.6	-	1.2	0.8
	63.9	57.1	5.6	8.5
VI D	62.1	57.5	1.7	1.2
	59.9	59.6	1.4	1.6

TABLE 70

Charpy Impact Results for Task II Cast Alloys

Alloy	Temperature R. T.	1400°F	1875°F
IV Y	35.0*	31.5	14.5
	40.0 (35.0)	34.0 (34.1)	15.0 (14.9)
	34.0	33.5	15.0
	31.0	33.5	15.0
VI A	38.0	32.0	9.5
	21.5 (28.2)	29.0 (31.4)	11.5 (12.0)
	18.5	37.5	13.0
	35.0	27.0	14.0
VI D	31.0	29.5	12.5
	22.0 (26.4)	19.0 (23.9)	15.0 (13.5)
	26.0	22.0	11.5
	28.5	25.0	15.0

* All data are in ft.-lb. Average value is given in parenthesis.

TABLE 71

Thermal Fatigue Results for Task II Cast Alloys

Alloy	First Cracking Observed	Failure	No Cracking Test Halted	Test Temperature, °F
IV Y				
1	4100 cycles	5100 cycles		2100
2			5100* cycles	2100
3			5300*	2100
4			5100*	2100
5			4000	1875
6			4000	1875
VI A				
1	1300	3600		2100
2	1400	2900		2100
3	1500	4100		2100
4	1900	3600		2100
5			4000	1875
6			4000	1875
VI D				
1	700	900		2100
2	200	600		2100
3	1400	2100		2100
4	900	2900		2100
5			4000	1875
6			4000	1875
Inco 713C (Virgin)				
1	1000	1400		2100
2	1150	1600		2100
IN 100 (Virgin)				
1	450	1600		2100
2	150	1600		2100
IN 100 (Revert)				
1	450	1050		2100
2	300	300		2100
WI 52 (Air Melt)				
1	900	1250		2100
2	200	350		2100

* Disqualified due to build up of residue on surface.

TABLE 72

Oxidation Results for Task II Cast Alloys

Alloy	Exposure Time, hours	Weight Gained in mg/cm^2 1875°F	1950°F	2100°F
IV Y	50	0.7	0.9	0.5
	200	1.2	1.6	0.9
	500	2.2	2.0	1.3 *
	1000	2.2	4.1	21.8 *
VI A	50	0.4	0.6	0.9
	200	0.7	1.2	5.0
	500	1.1	1.8	30.5 *
	1000	1.4	3.2	11.4
VI D	50	6.9	4.1	3.9
	200	36	40	322.5 **
	500	232 **	509 **	528
	1000	407 *	519	477

* Surface melting observed.
** Catastrophic oxidation had occurred by this time.

TABLE 73

Hot Corrosion Results for Task II Cast Alloys

Alloy	Weight Loss (grams) *
IV Y	0.529
VI A	0.150
	0.271
VI D	1.633
	1.968

* Weight loss after 1 hour exposure to 1% NaCl, 99% Na_2SO_4 and 1800°F.

TABLE 74

Phases Identified by X-Ray Diffraction of Extracted Residues of Task II Cast Alloys

Alloy	Condition	Phase *			
		MC	$M_{23}C_6$	M_6C	M_3B_2
IV Y	As-Cast	VS		? VW	VW
	Aged at 1600°F/1500 hrs.	VS	W	? VW	W
VI A	As-Cast	S		? VW	W
	Aged at 1600°F/1500 hrs.	VS			W
	Failed in S.R. test at 1875°F/ 15,000 psi (1174.3 hrs.)	S			VW
	Aged at 1875°F/300 hrs. in argon	S		W	VW
VI D	As-Cast	S	VW	VW	W
	Aged at 1600°F/1500 hrs.	VS		VW	W

* Residues extracted electrolytically in a 10% HCl methanol solution.
S represents strong; M, medium; W, weak; and V, very.

TABLE 75

Stress Rupture Results for Task II Wrought Alloys Tested in Air Atm. at 1875°F

	10,000 psi			15,000 psi		
Alloy	Life (hours)	Elongation (%)	R. A. (%)	Life (hours)	Elongation (%)	R. A. (%)
I-5	227.3	8.0	6.2	46.0	4.3	1.8
	327.3	10.1	6.8	30.3	4.1	1.9
	285.0	7.7	8.0	43.0	5.1	5.6
IIb	263.2	4.9	3.7	40.4	3.4	0.6
	251.2	7.8	2.5	38.5	4.2	2.4
	231.6	5.7	4.3	50.3	4.6	2.5
IId	298.7	8.1	2.5	23.6	4.7	1.9
	221.6	3.9	3.1	45.4	2.8	1.9
	34.5*	-	-	41.0	14.2	36.2
IIIg	252.3	17.6	19.6	72.3	9.7	9.2
	386.4	17.6	43.0	69.4	8.5	10.9
	326.0	20.4	24.5	78.2	7.2	10.9
IV Y	3.2	-	7.4	160.8**	2.6	0.6
	6.95***	-	-	35.9	1.8	0.6
	0.09	0.3	0	177.5***	-	-
U 700	100					

* Result is no good. Material reacted with grip material.
** 5000 psi load for IV Y specimens.
*** Broke in threads.

Heat Treatments: Alloys I-5, IIb, IId, and IIIg - 2250/2/AC + 2000/4/AC + 1550/24/AC + 1400/16/AC
Alloy IV Y - 2290/2/AC + 2000/4/AC + 1550/24/AC + 1400/16/AC

TABLE 76

Average Tensile Results for Nickel-Base Cast Alloys

		Room Temperature				1200°F			
Alloy	Condition	Ultimate (10^3psi)	0.2% Offset Yield(10^3psi)	Elongation (%)	R.A. (%)	Ultimate (10^3psi)	0.2% Offset Yield(10^3psi)	Elongation (%)	R.A. (%)
IV Y	As-Cast*	138	126	3.6	8.5	160	133	5.8	10.2
	Argon Aged	126	115	1.6	3.4	127	112	2.5	5.0
VI A	As-Cast *	152	136	4.2	6.0	165	137	3.9	7.0
	Argon Aged	131	116	2.8	4.1	133	112	3.7	7.4
VI D	As-Cast *	146	132	2.9	5.6	154	137	2.2	3.2
	Argon Aged	115	107	2.4	6.2	131	118	2.1	5.5
Inco 713C[23]	As-Cast	123	107	7.9					
MAR M-200[23]	As-Cast	135	120	7.0					
IN 100[23]	As-Cast	147	123	9.0					
TAZ 8[24]	As-Cast	134							
TAZ 8A[25]	As-Cast	128		5					

* Average Results include Task I and Task II results.

TABLE 76 (Continued)

Average Tensile Results for Nickel-Base Cast Alloys

		1400°F				1600°F			
Alloy	Condition	Ultimate (10^3psi)	0.2% Offset Yield(10^3psi)	Elongation (%)	R.A. (%)	Ultimate (10^3psi)	0.2% Offset Yield(10^3psi)	Elongation (%)	R.A. (%)
IV Y	As-Cast	156	133	5.1	8.8	121	115	6.0	7.8
	Argon Aged	136	127	1.6	3.0	108	-	1.2	1.6
VI A	As-Cast	159	137	4.4	5.5	126	112	2.5	4.1
	Argon Aged	123	115	2.2	4.4	110	-	2.0	1.6
VI D	As-Cast	151	136	3.2	5.5	126	115	2.8	4.1
	Argon Aged	126	121	1.2	3.7	103	-	1.3	2.4
Inco 713C	As-Cast	136	108	5.9		105	72	14	
MAR M-200	As-Cast	135	122	3.4		122	110	3.8	
IN 100	As-Cast	155	125	6.5		128	101	6.0	
TAZ 8	As-Cast								
TAZ 8A	As-Cast	129		2.5		109		3.7	

TABLE 76 (Continued)

Average Tensile Results for Nickel-Base Cast Alloys

Alloy	Condition	1875°F Ultimate (10^3 psi)	1875°F 0.2% Offset Yield (10^3 psi)	1875°F Elongation (%)	1875°F R. A. (%)	1950°F Ultimate (10^3 psi)	1950°F 0.2% Offset Yield (10^3 psi)	1950°F Elongation (%)	1950°F R. A. (%)
IV Y	As-Cast	70	62	4.9	5.2	58	52	5.8	7.2
	Argon Aged	63	56	3.2	3.9				
VI A	As-Cast	71	61	5.5	6.6	57	50	4.9	4.3
	Argon Aged	58	-	3.4	4.6				
VI D	As-Cast	72	64	4.6	7.8	61	55	2.2	2.2
	Argon Aged	62	59	1.6	1.4				
		1800°F				1900°F			
Inco 713C	As-Cast	68	44	20					
MAR M200	As-Cast	80	68	4.5					
IN 100	As-Cast	82	54	6.0					
TAZ-8	As-Cast	80				56			
TAZ 8A	As-Cast	80		5.5		64		4.0	

TABLE 76 (Continued)

Average Tensile Results for Nickel-Base Cast Alloys

		2000°F			
Alloy	Condition	Ultimate (10^3 psi)	0.2% Offset Yield (10^3psi)	Elongation (%)	R. A. (%)
IV Y	As-Cast	48	46	4.0	3.5
VI A	As-Cast	49	45	4.8	5.7
VI D	As-Cast	49	45	7.8	7.6
Inco 713C	As-Cast				
MAR M200	As-Cast	47			
IN 100	As-Cast	42			
TAZ 8	As-Cast	49			
TAZ 8A	As-Cast	49		4.6	

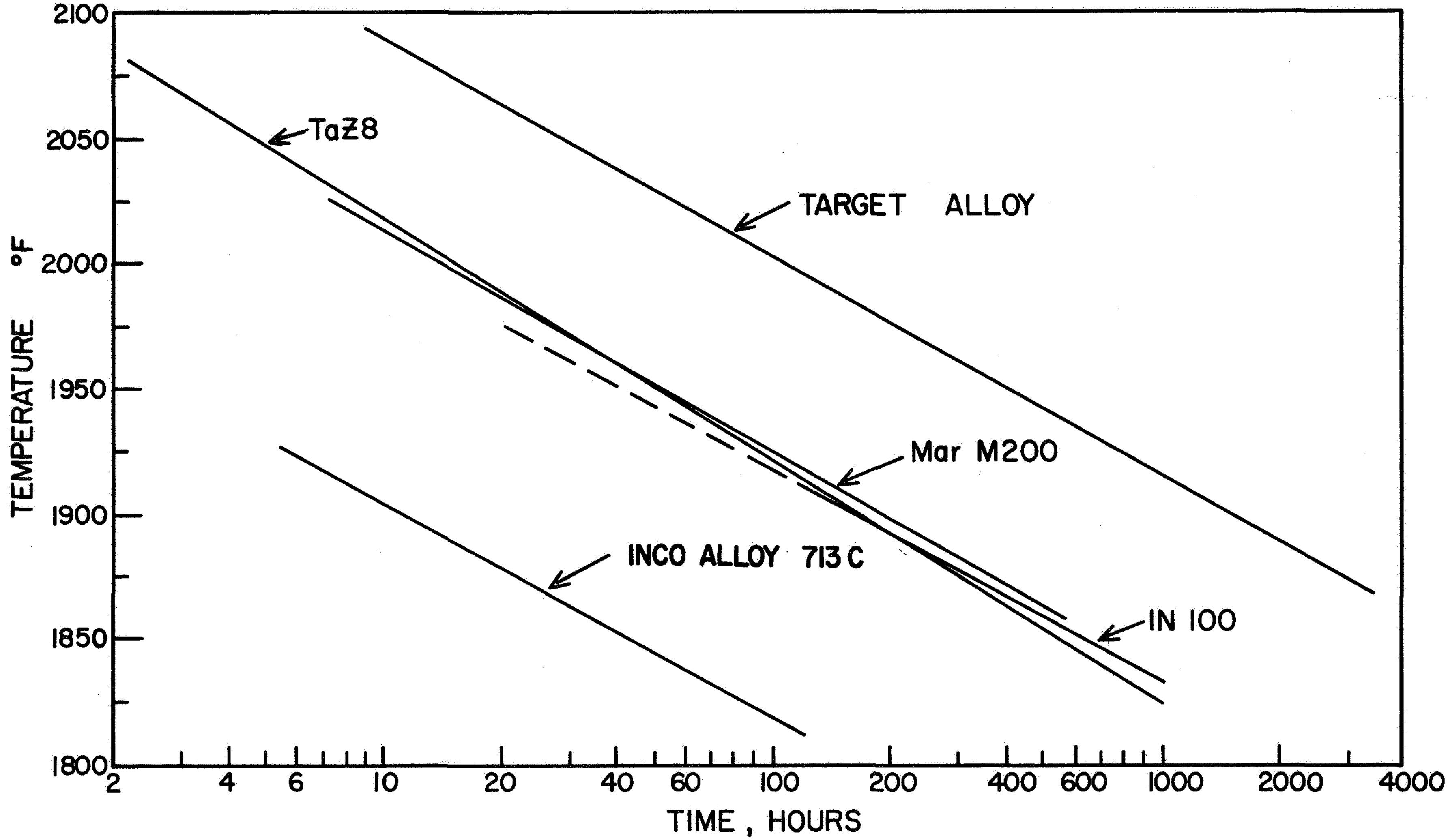
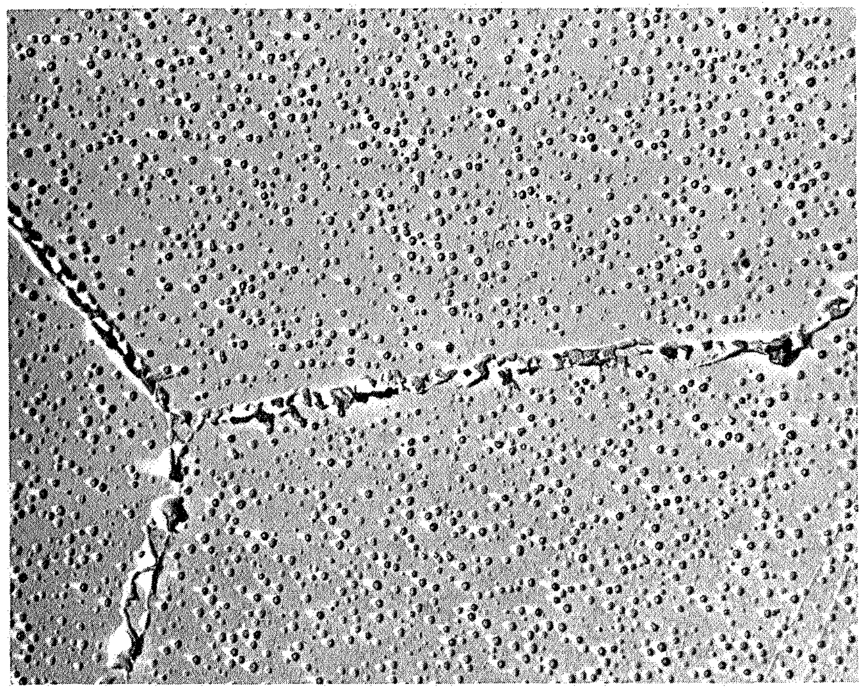
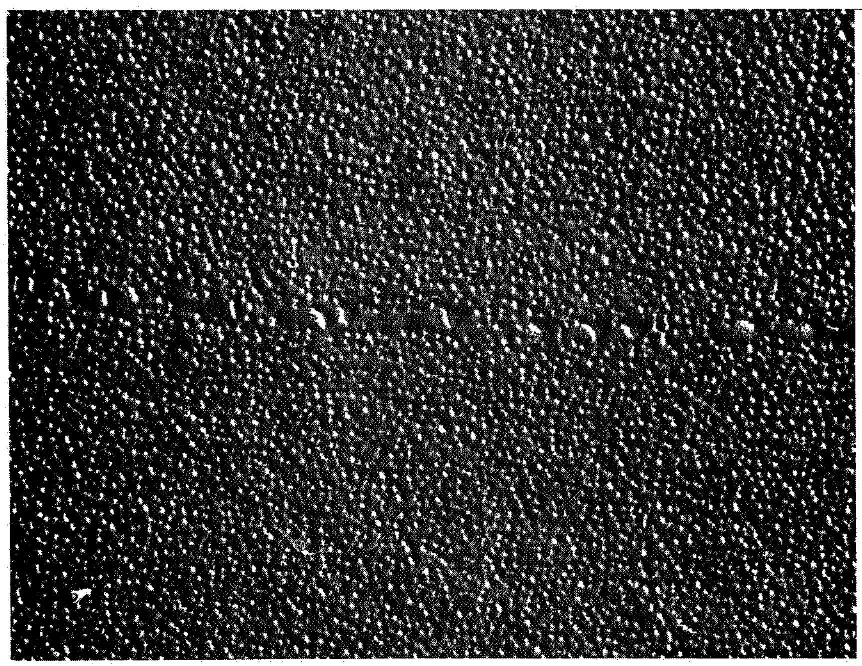


Figure 1. Stress Rupture Properties at 15,000 psi of Existent Cast Nickel-Base Superalloys in Relation to Target Alloy. (After Freche and Waters)[3].

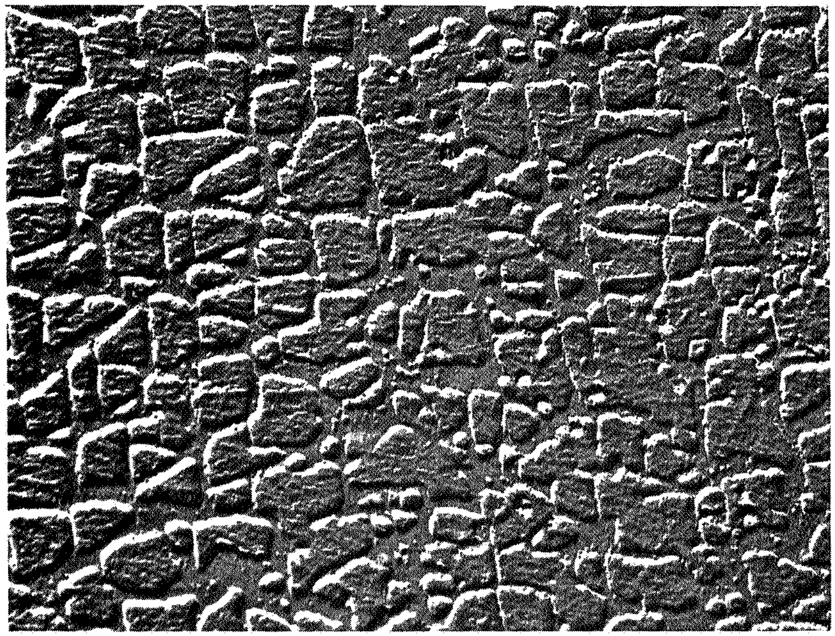


A. Waspaloy - Solution Treated at 2100°F, Aged at 1700°F. 6000X Magnification.

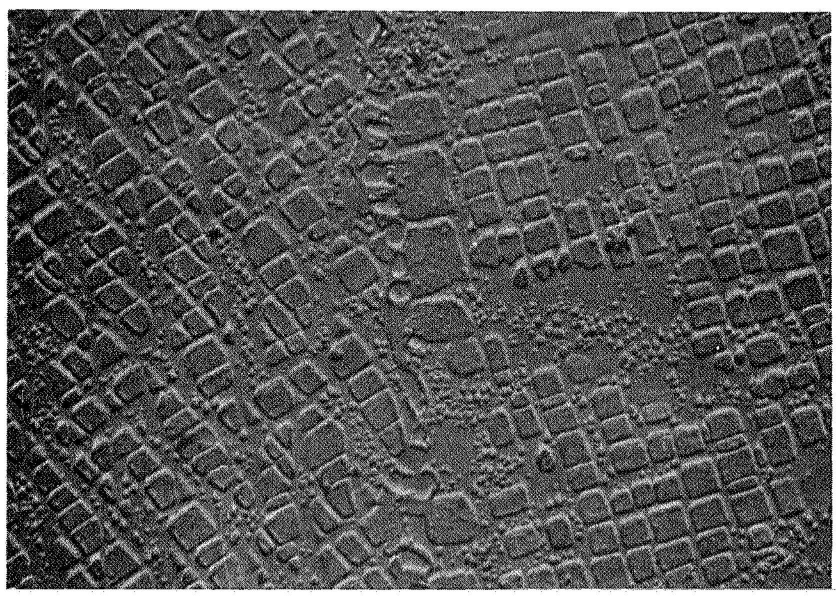


B. Udimet 700 - Solution Treated at 2150°F, Water Quenched. 15,000X Magnification.

Figure 2. Examples of Gamma-Prime Formations in Nickel-Base Superalloys.



A. Gamma Prime Formations in Cast TRW 1900.



B. Coarse and Fine Gamma Prime Formations in Heat Treated Nimonic 115.

Figure 3. Blocky Intragranular Gamma-Prime Formations in Nickel-Base Superalloys. 10,000X Magnification.

9873

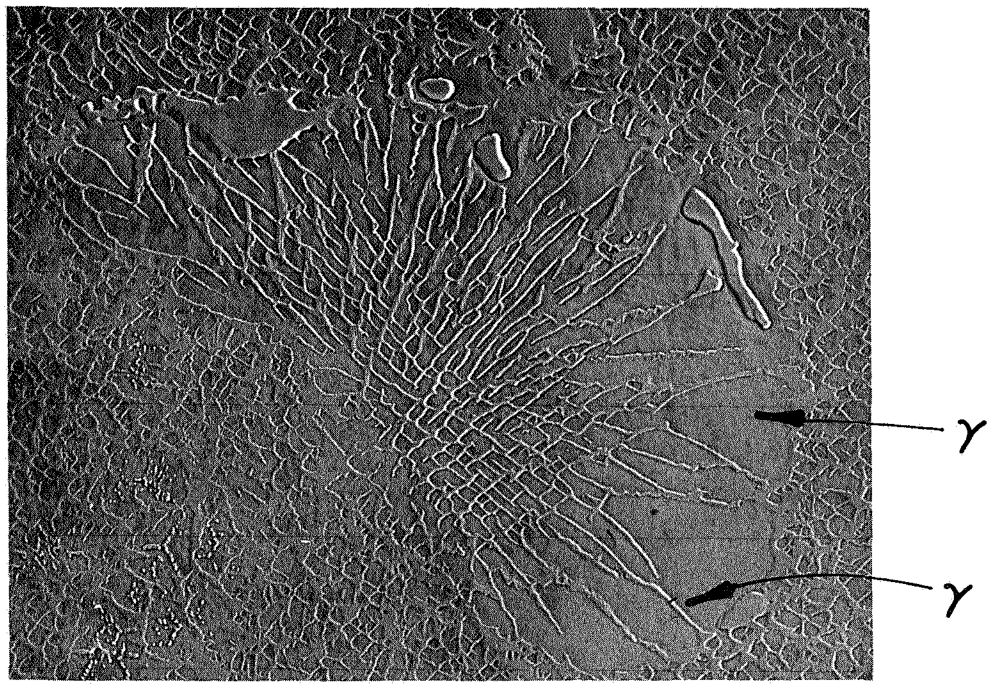
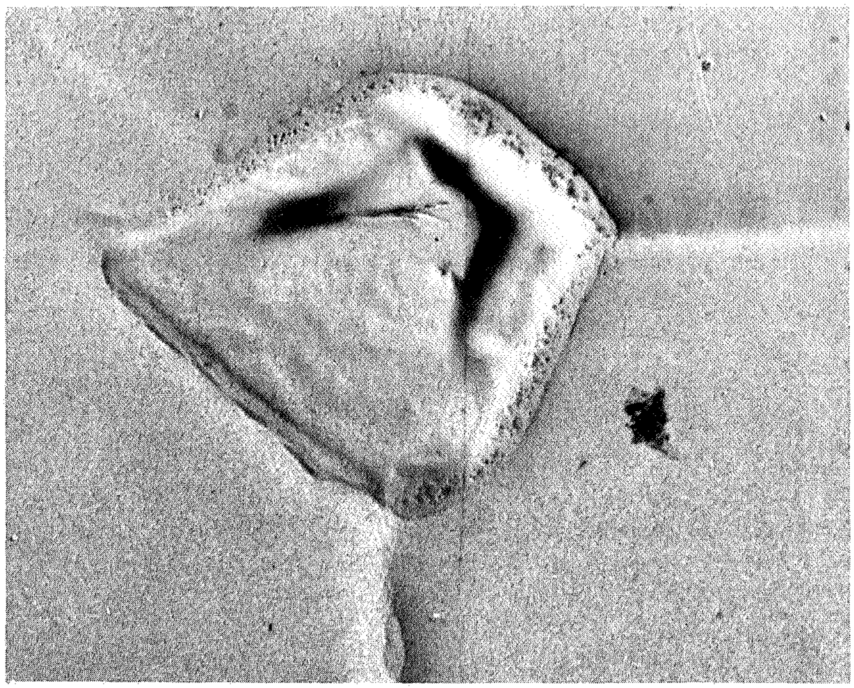
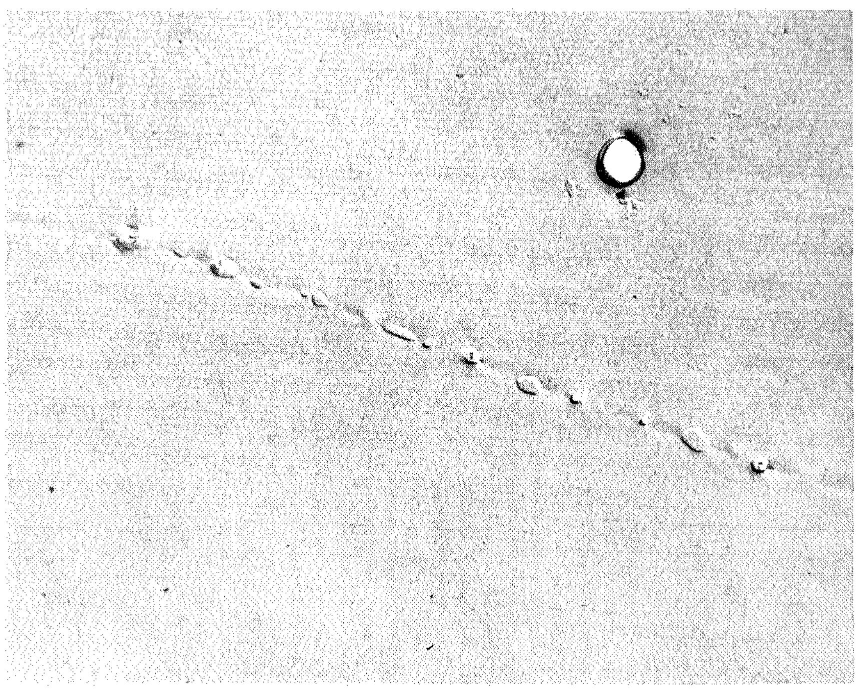


Figure 4. Eutectic Structure of Gamma Prime and Gamma Phases ("Primary Gamma Prime") in Cast Nickel-Base Superalloy TRW 1800. 3400X Magnification.

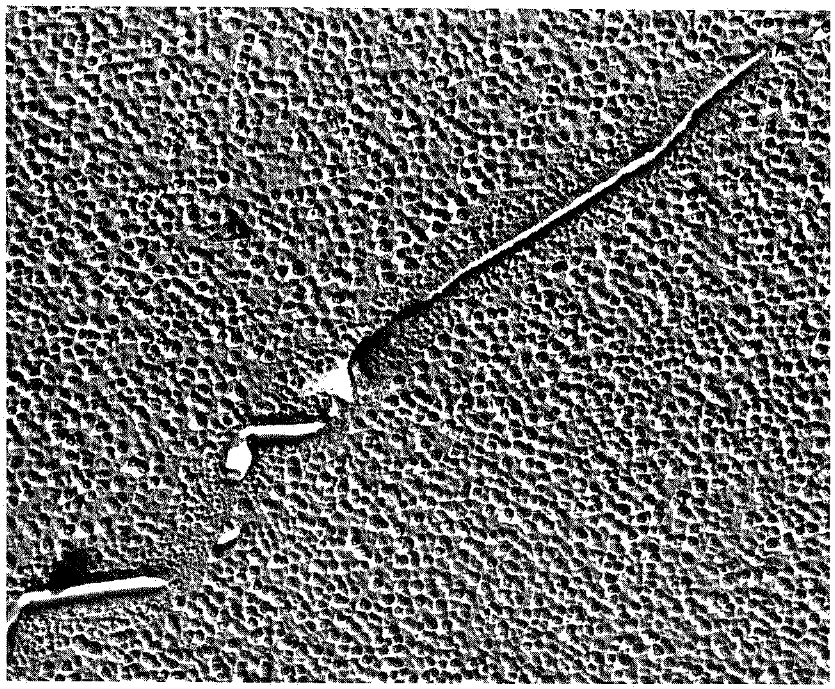


A. Massive MC Carbide Particle. 10,000 Magnification

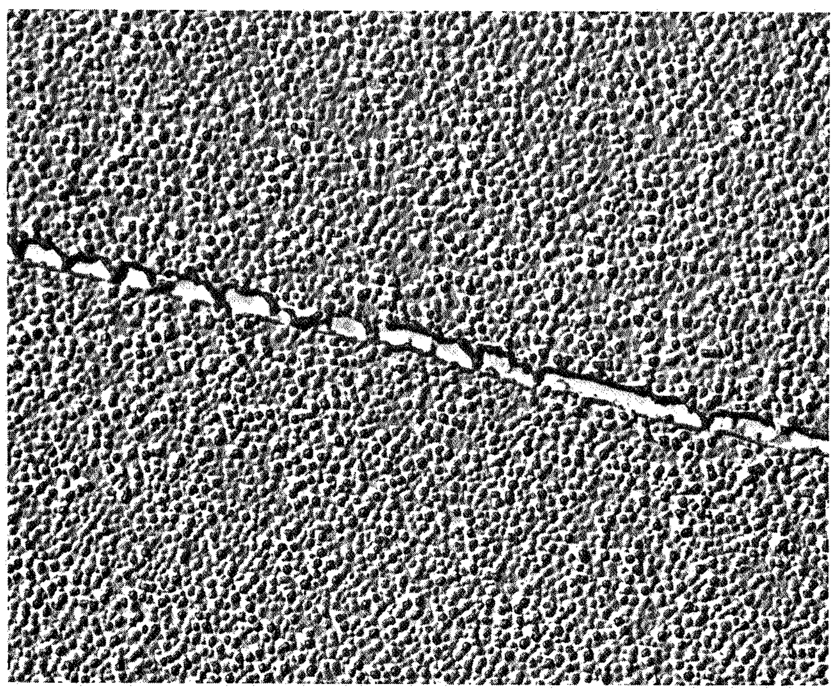


B. Grain Boundary MC Particles. 6000X Magnification

Figure 5. MC Carbide Formations in Waspaloy.



A) Platelet $M_{23}C_6$ Formation



B) Discrete Particles $M_{23}C_6$ Formation

Figure 6. Grain Boundary $M_{23}C_6$ Formation in Inconel 700. 15,000X Magnification.

1340A

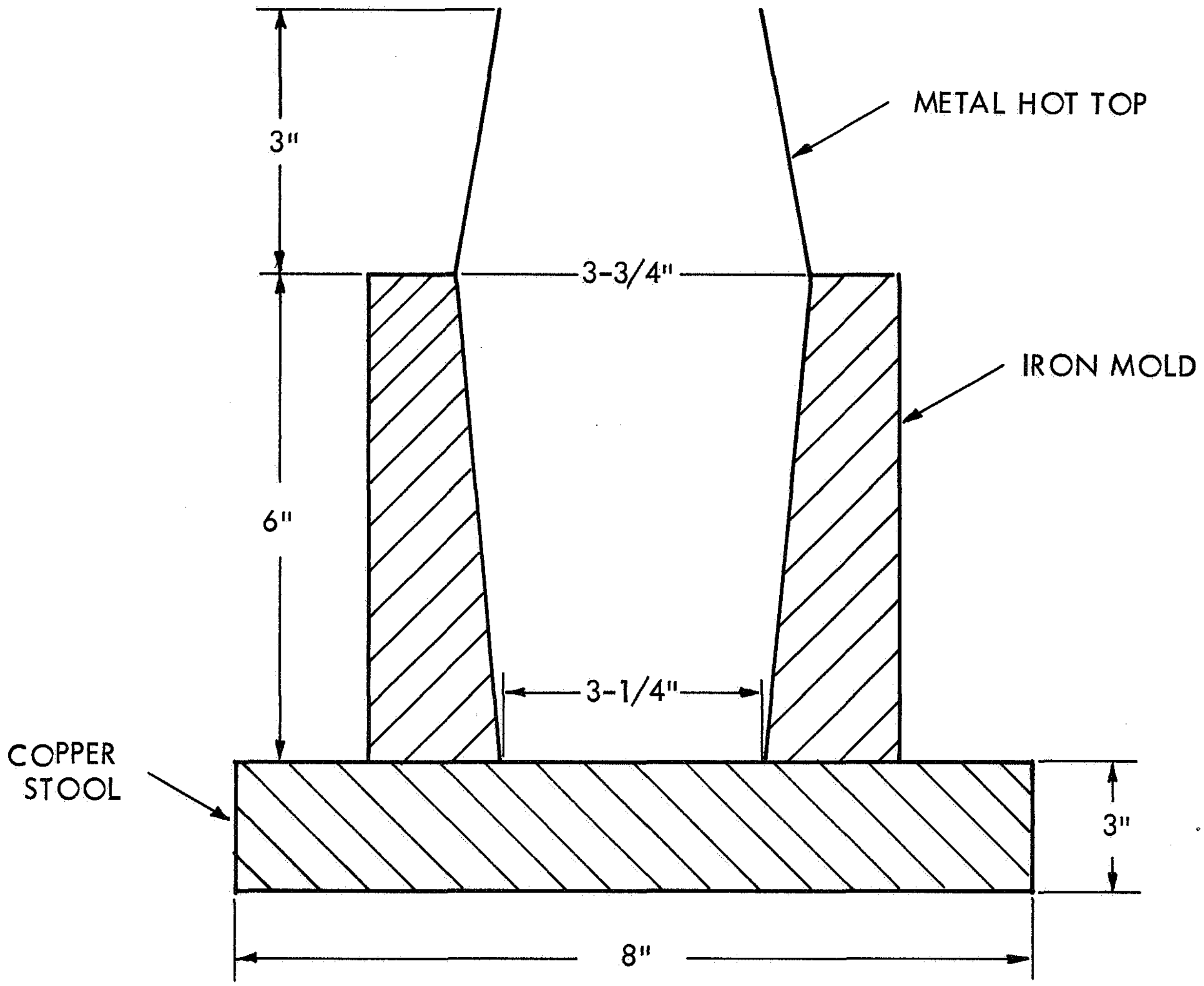


Figure 7. "Flower Pot" Type Mold Used in Casting Extrusion Ingots.

4271A

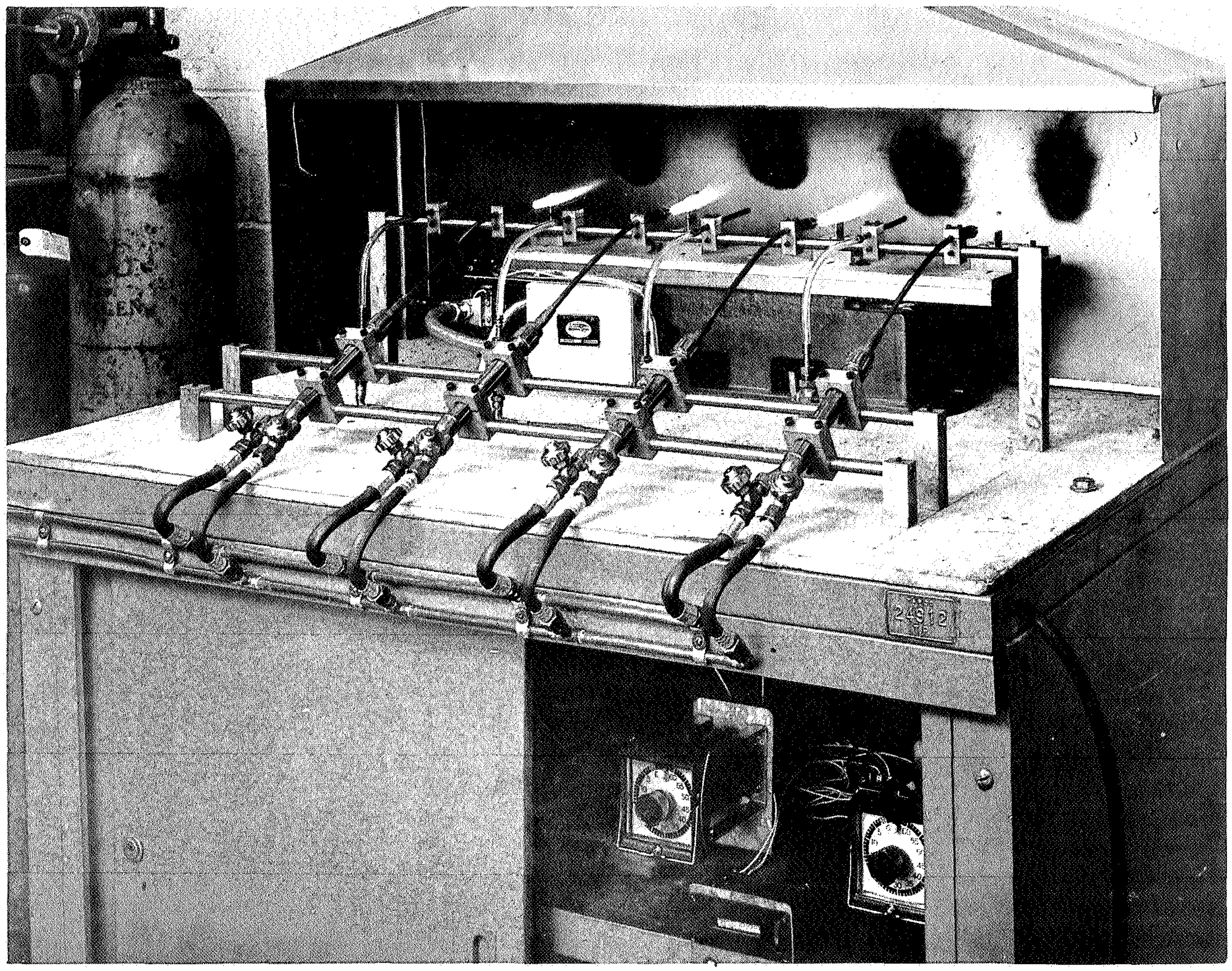


Figure 8. Thermal Fatigue Testing Apparatus.

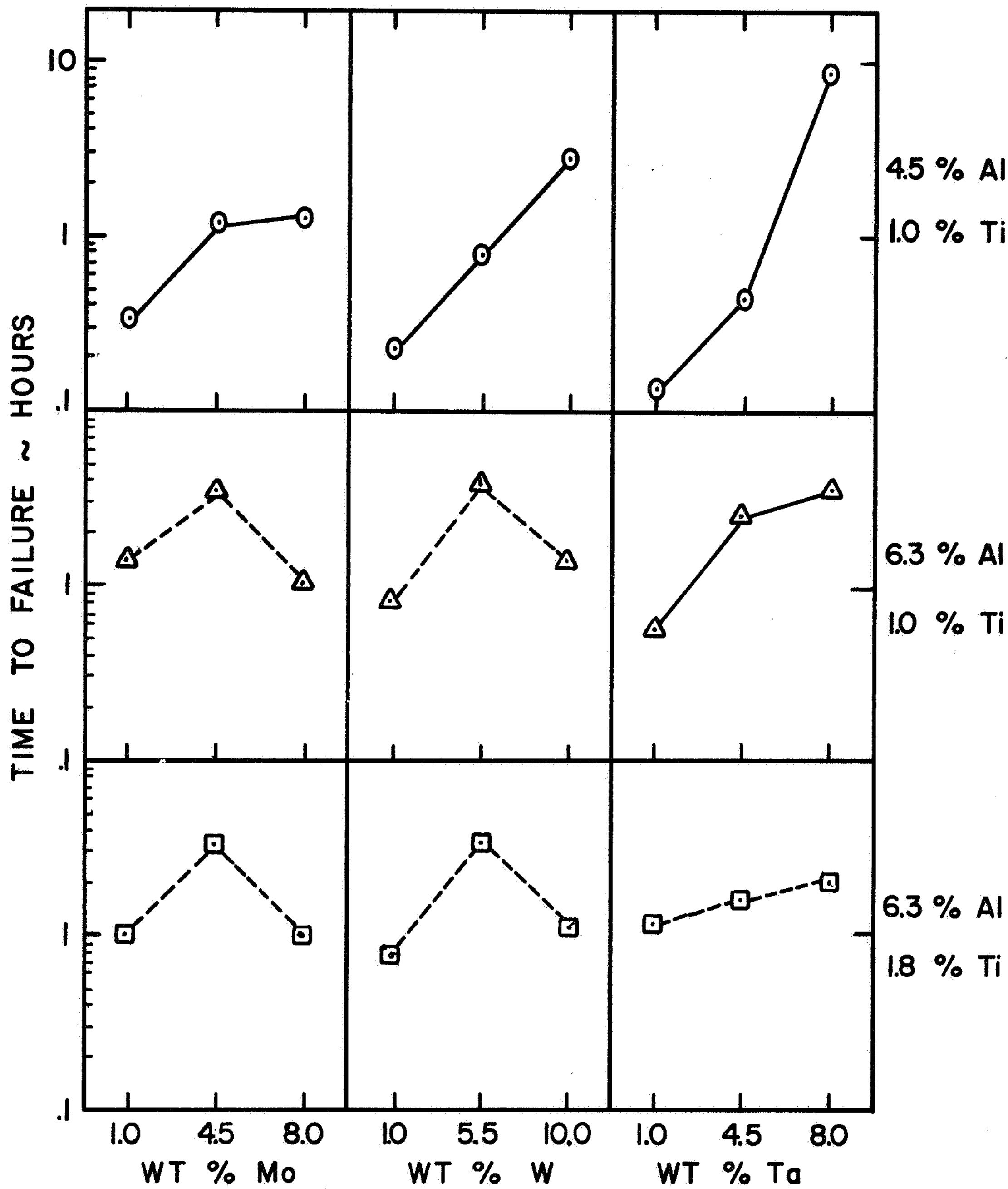


Figure 9. Average 2000°F, 15,000 psi Stress Rupture Life for Nickel-Base Alloys of Series I. Note that a factor of 10^2 was used for ease of Calculation for Series I Alloys. (Solid Lines are Statistically Significant.)

9862

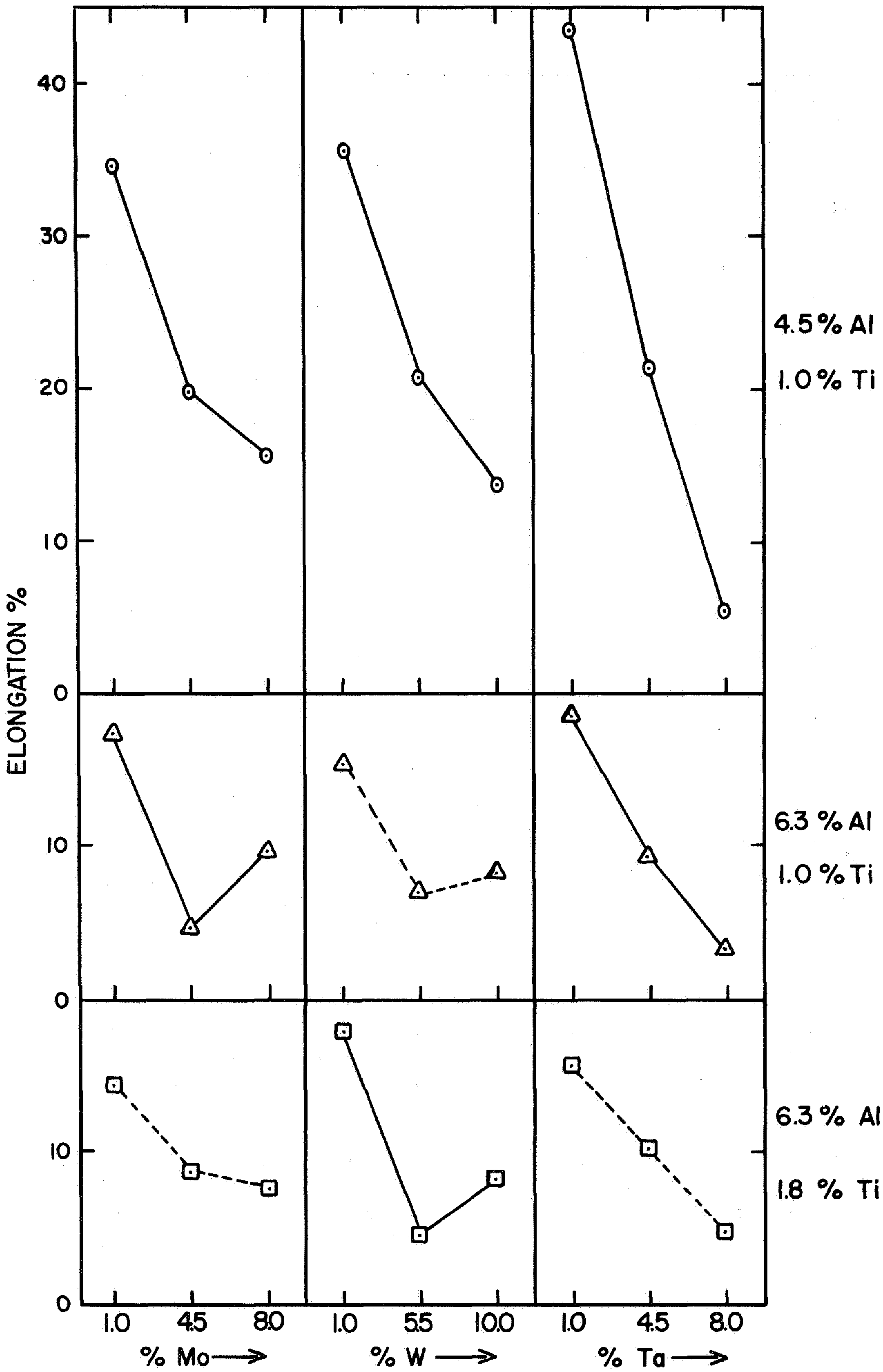


Figure 10. Average 2000°F, 15,000 psi Stress Rupture Elongation for Nickel-Base Alloys of Series I. (Solid Lines are Statistically Significant.)

9876

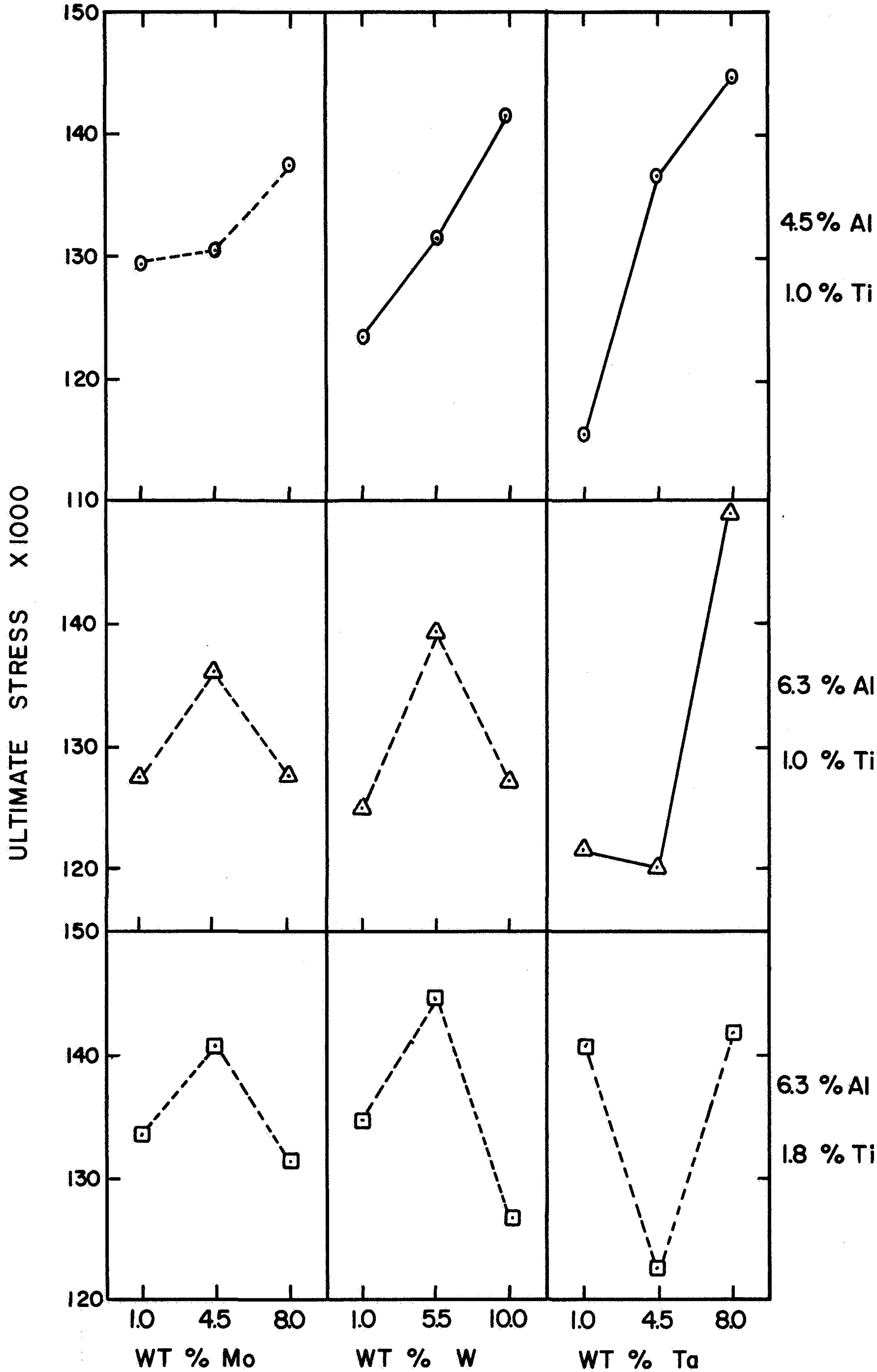


Figure 11. Average Room Temperature Tensile Strength for Nickel-Base Alloys of Series I. (Solid Lines are Statistically Significant.)

9875

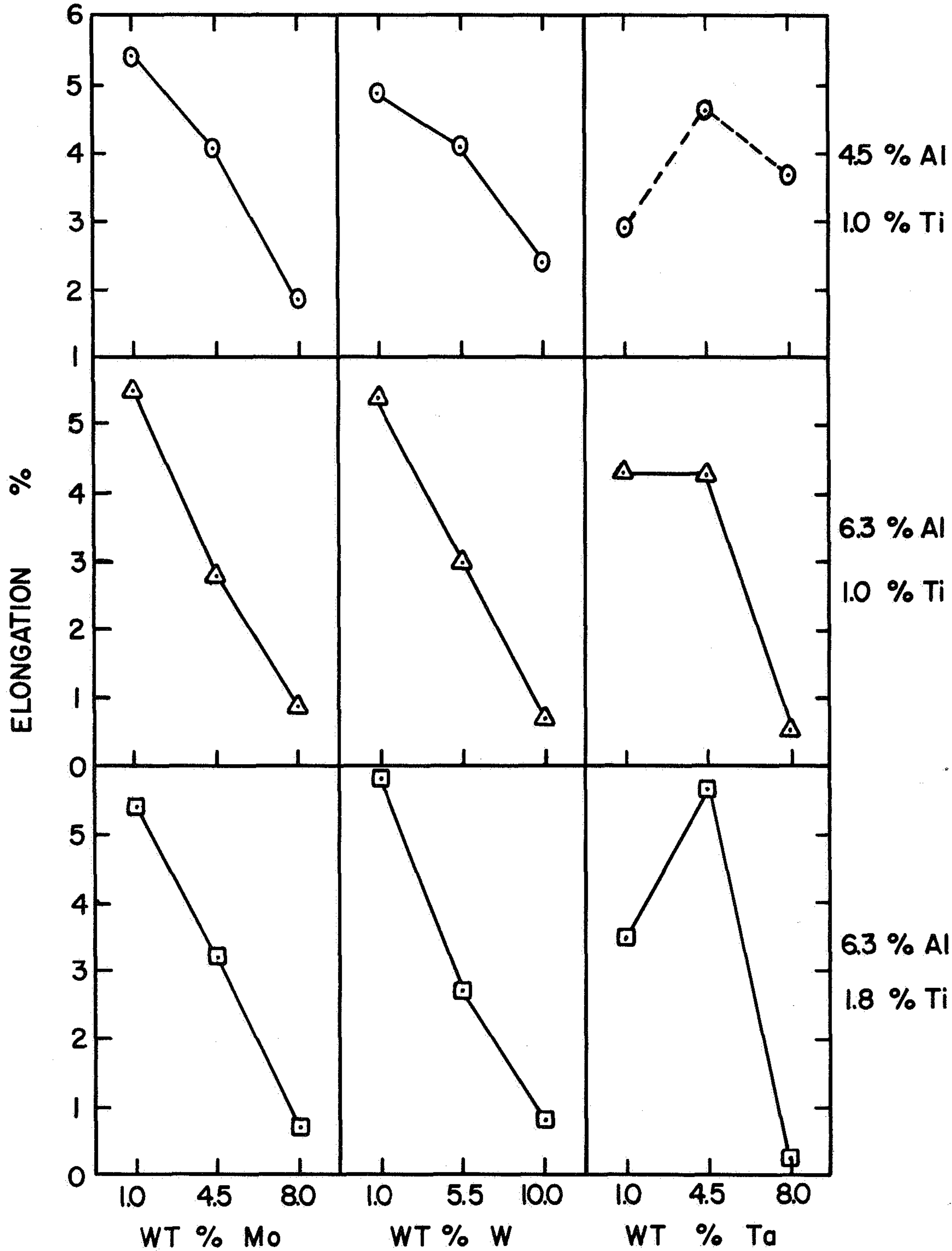


Figure 12. Average Room Temperature Tensile Elongation for Nickel-Base Alloys of Series I. (Solid Lines are Statistically Significant.)

9868

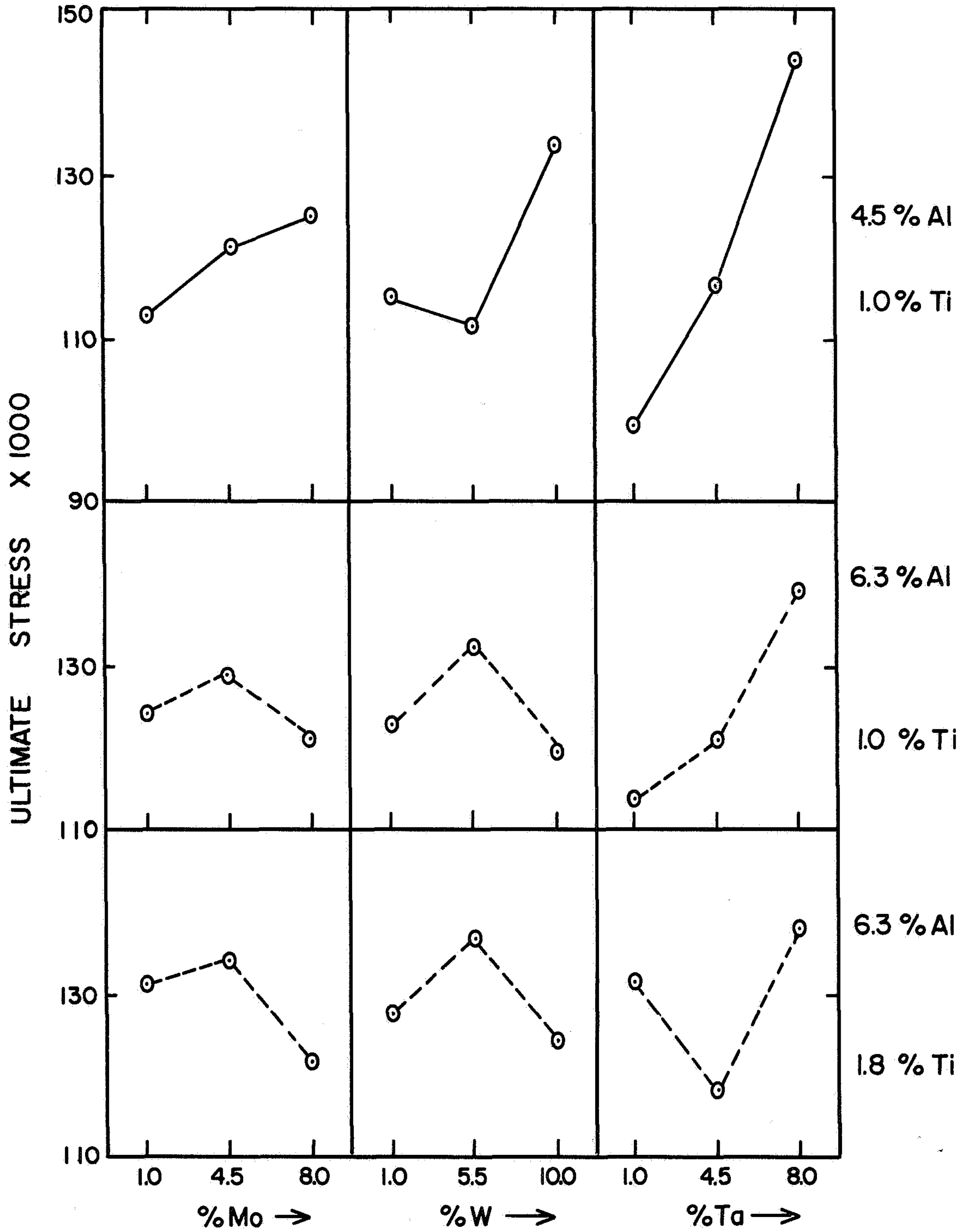
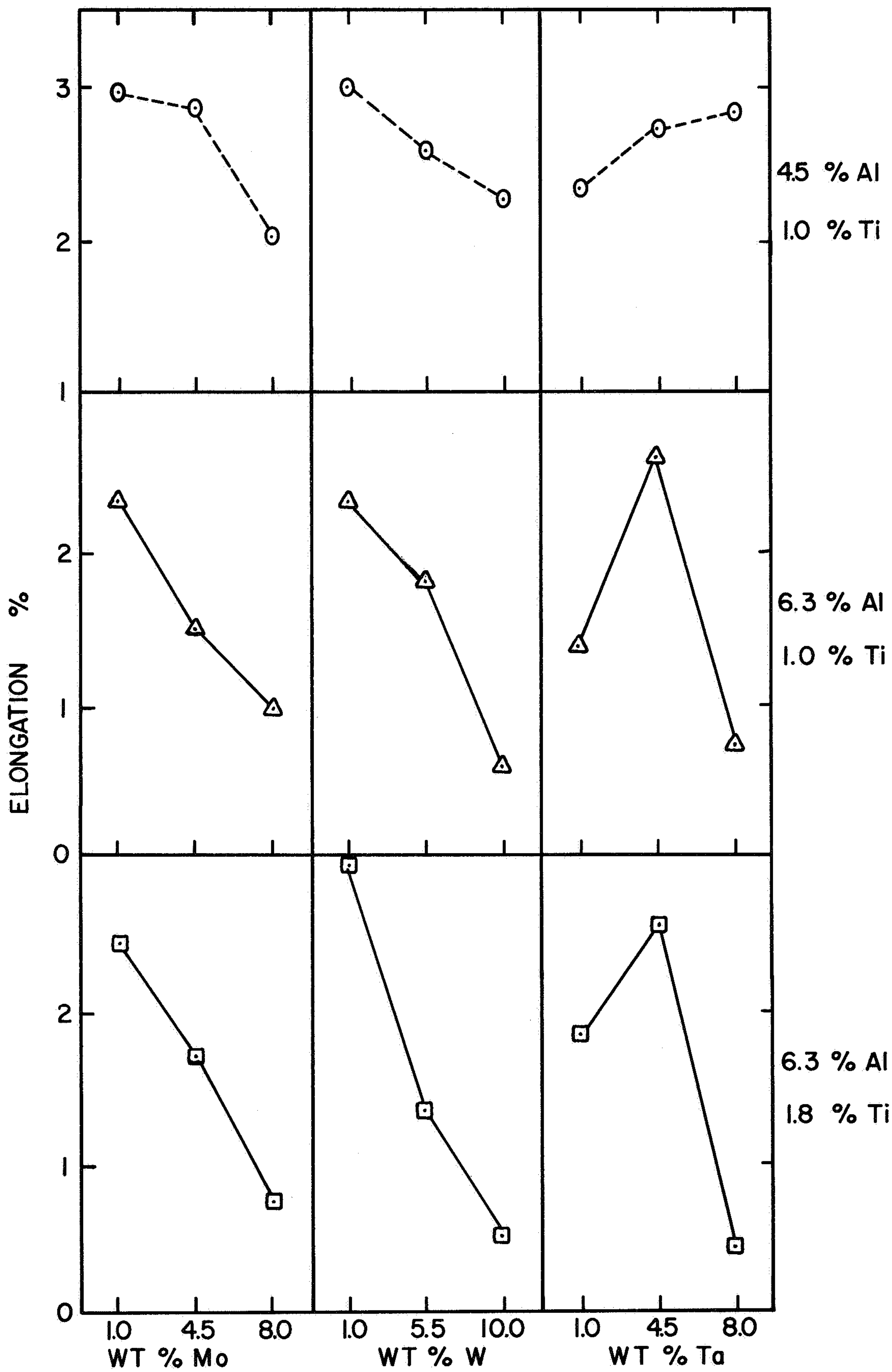


Figure 13. Average 1400°F Tensile Strength for Nickel-Base Alloys of Series I. (Solid Lines are Statistically Significant.)

9866



9864

Figure 14. Average 1400°F Tensile Elongation for Nickel-Base Alloys of Series I. (Solid Lines are Statistically Significant.)

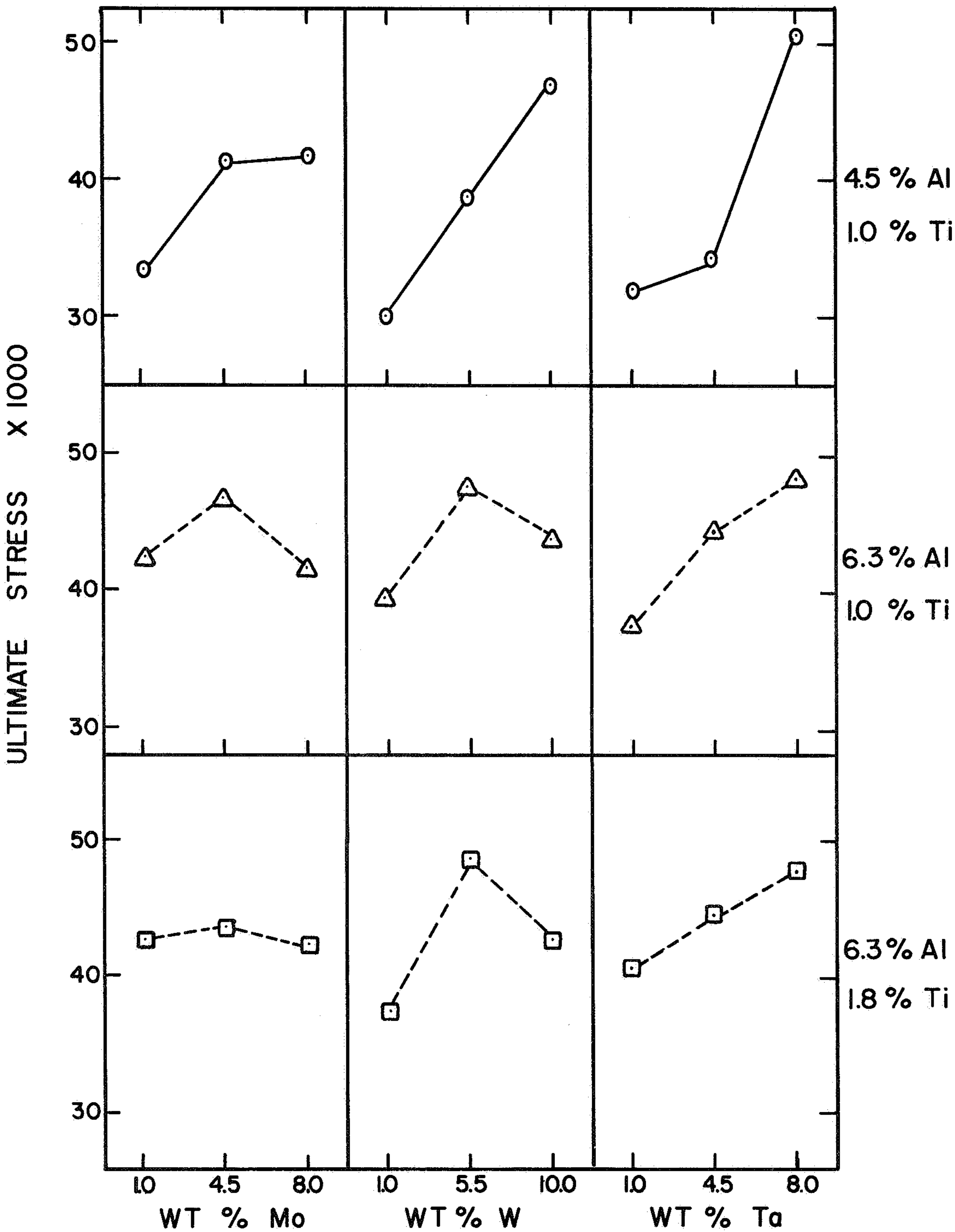


Figure 15. Average 1875°F Tensile Strength for Nickel-Base Alloys of Series I. (Solid Lines are Statistically Significant.)

9861

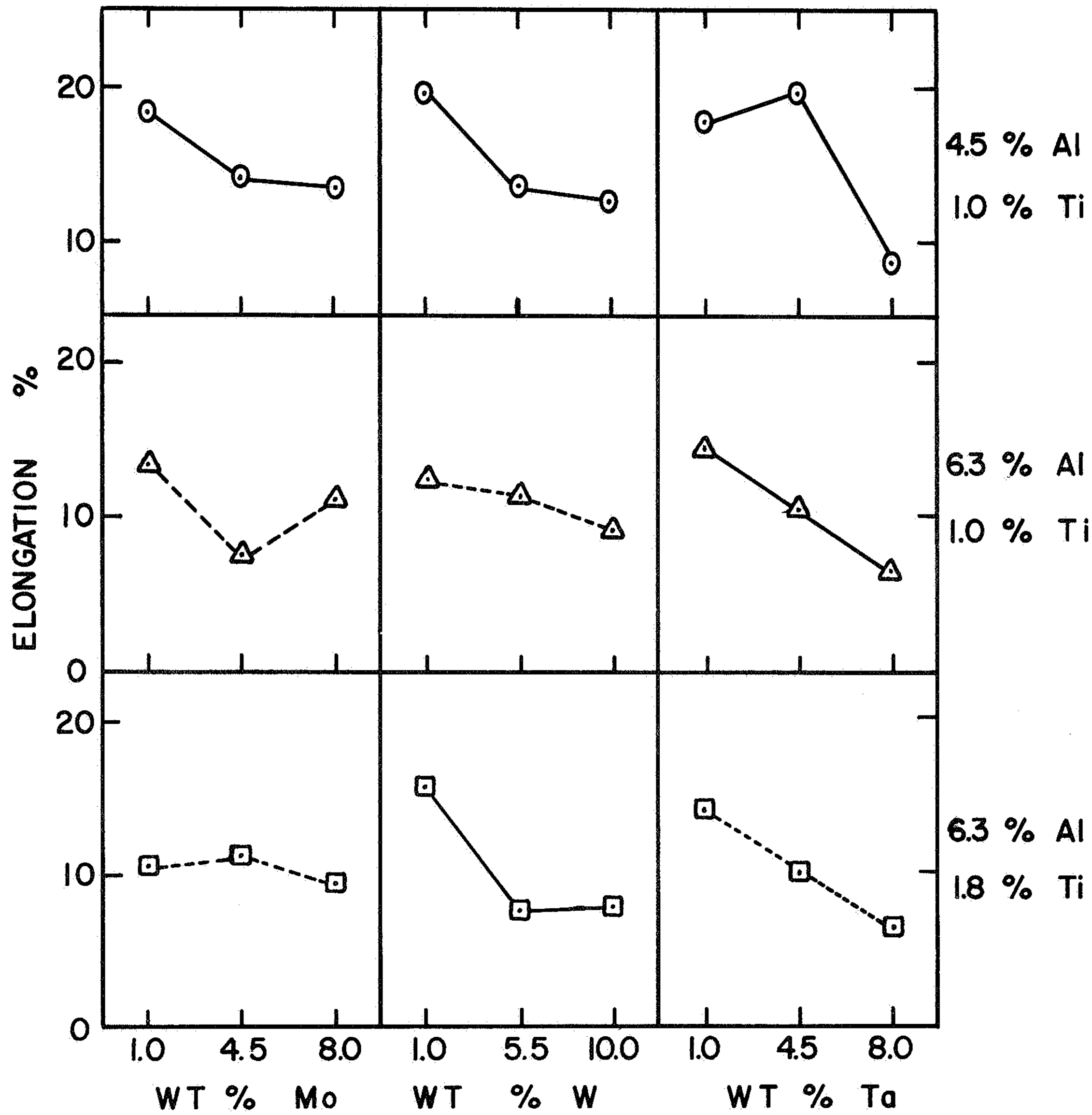


Figure 16. Average 1875°F Tensile Elongation for Nickel-Base Alloys of Series I. (Solid Lines are Statistically Significant.)

9867

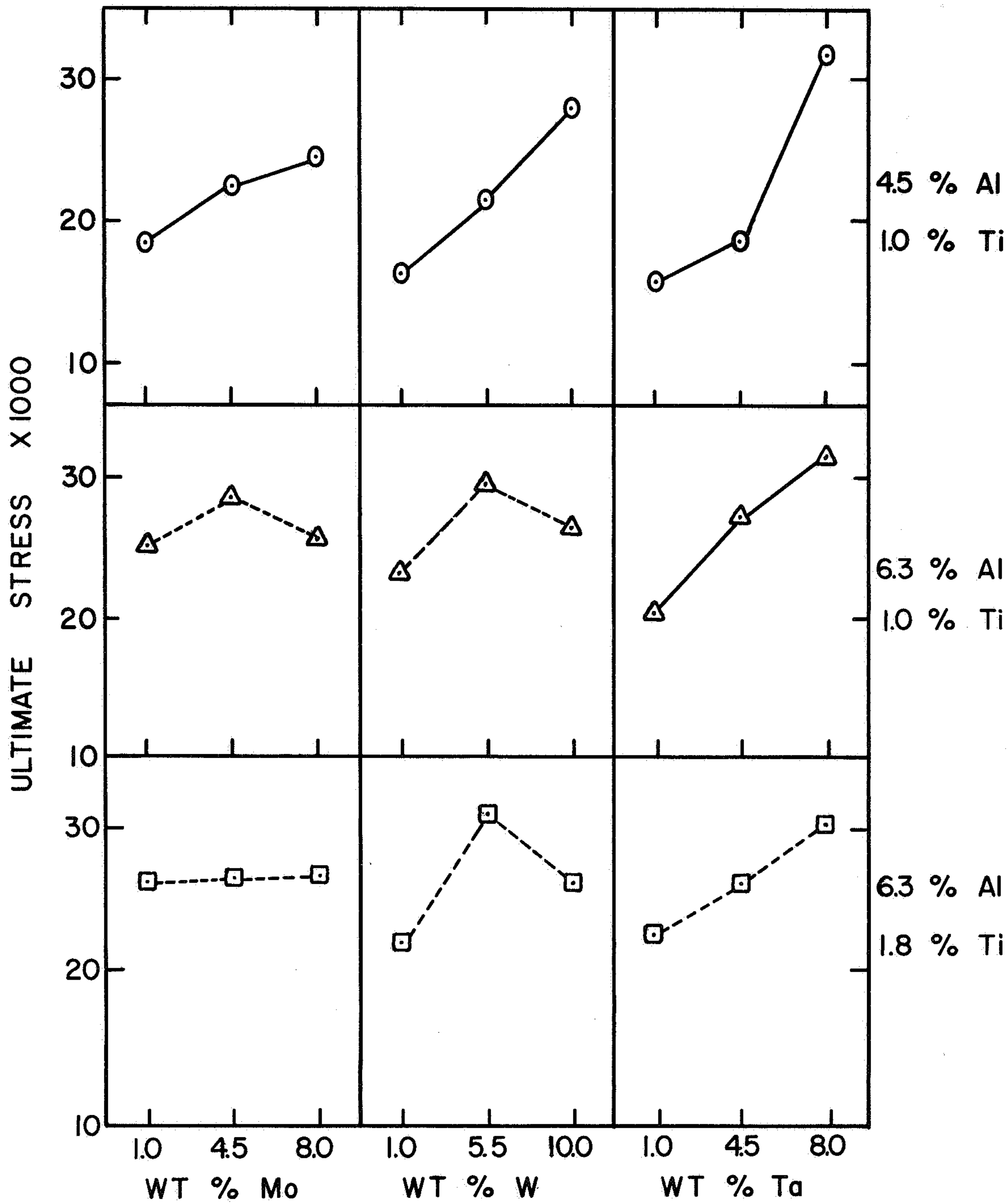


Figure 17. Average 2000°F Tensile Strength for Nickel-Base Alloys of Series I. (Solid Lines are Statistically Significant.)

9859

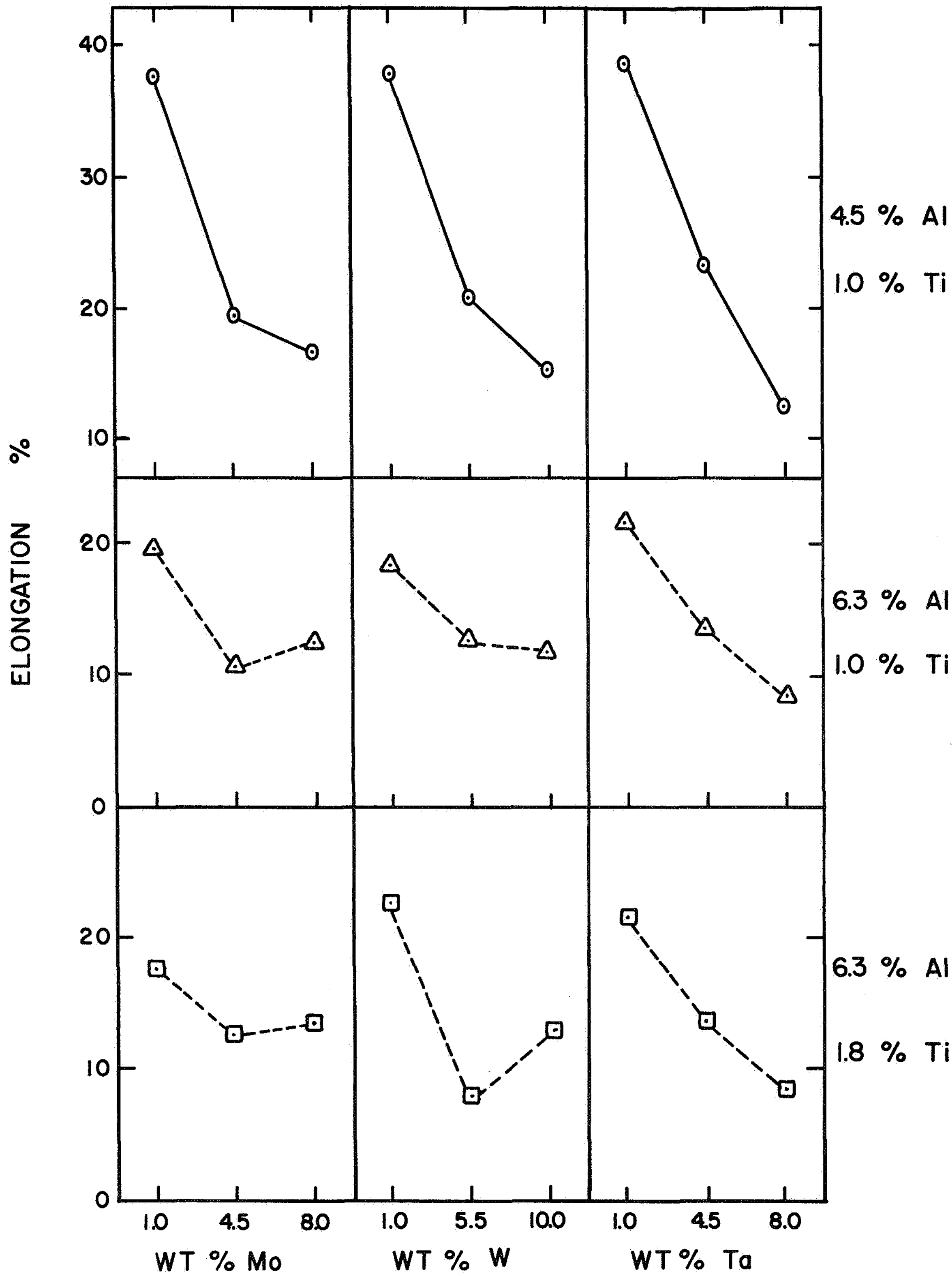


Figure 18. Average 2000°F Tensile Elongation for Nickel-Base Alloys of Series I. (Solid Lines are Statistically Significant.)

9865

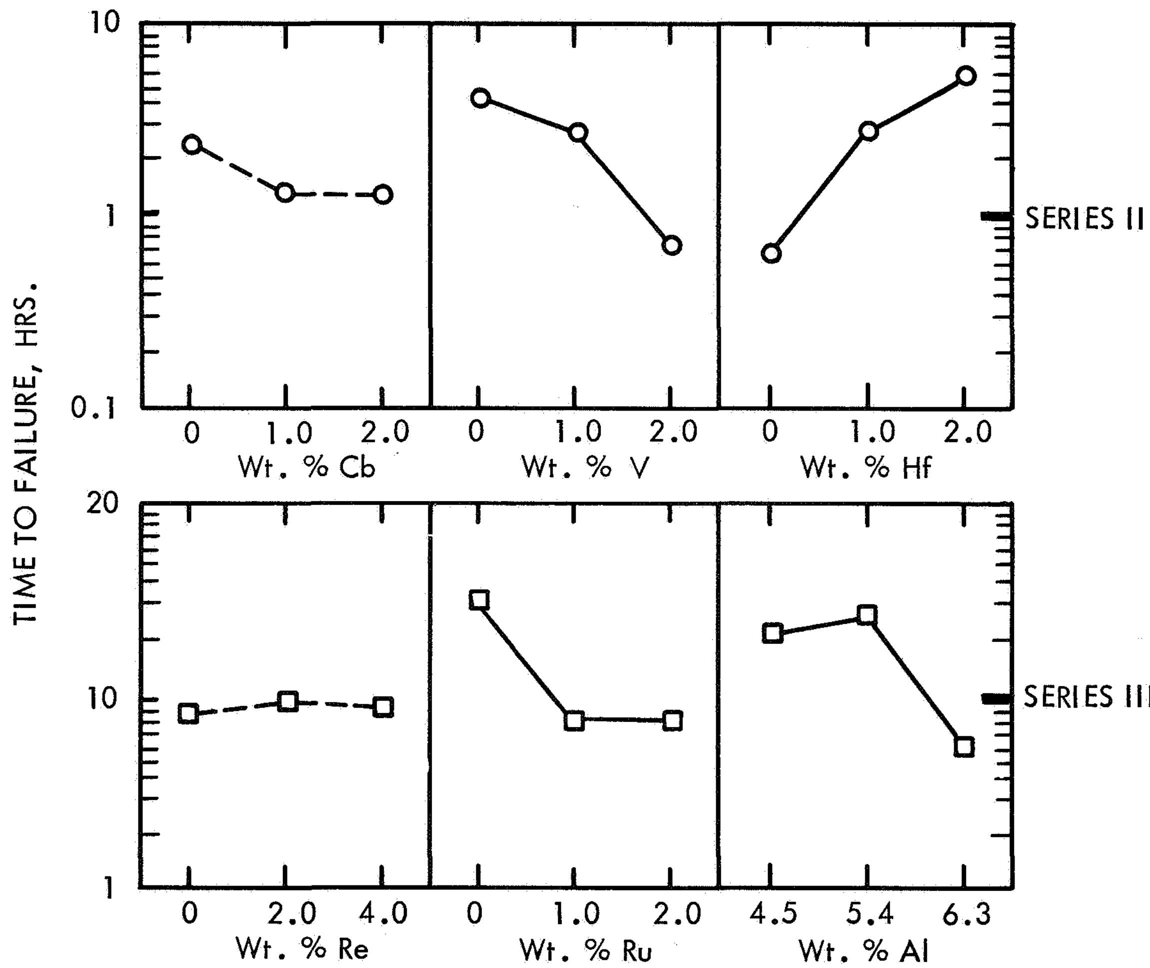


Figure 19. Average 2000°F, 15,000 psi Stress Rupture Life for Nickel-Base Alloys of Series II and III. (Solid Lines are Statistically Significant.)

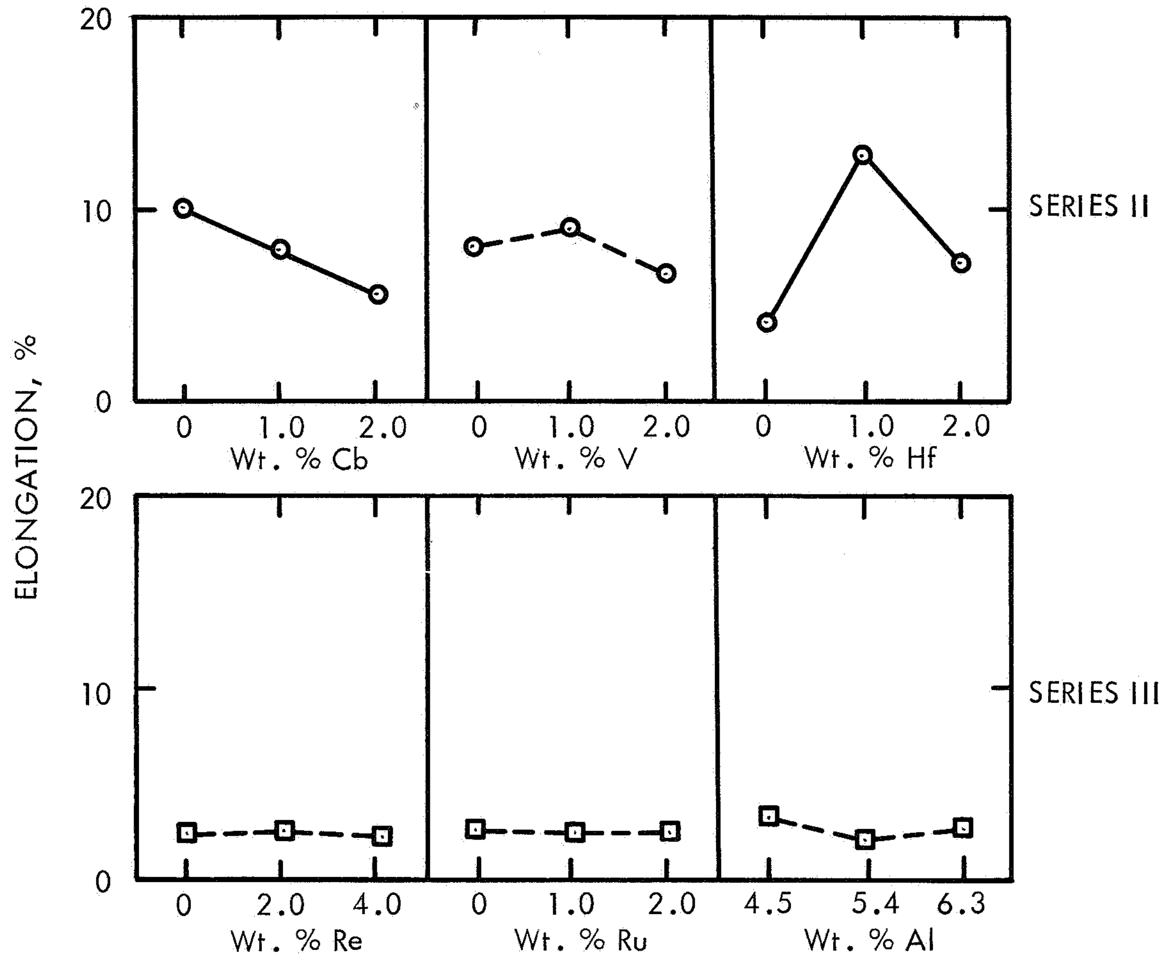


Figure 20. Average 2000°F, 15,000 psi Stress Rupture Elongation for Nickel-Base Alloys of Series II and III. (Solid Lines are Statistically Significant.)

1975A

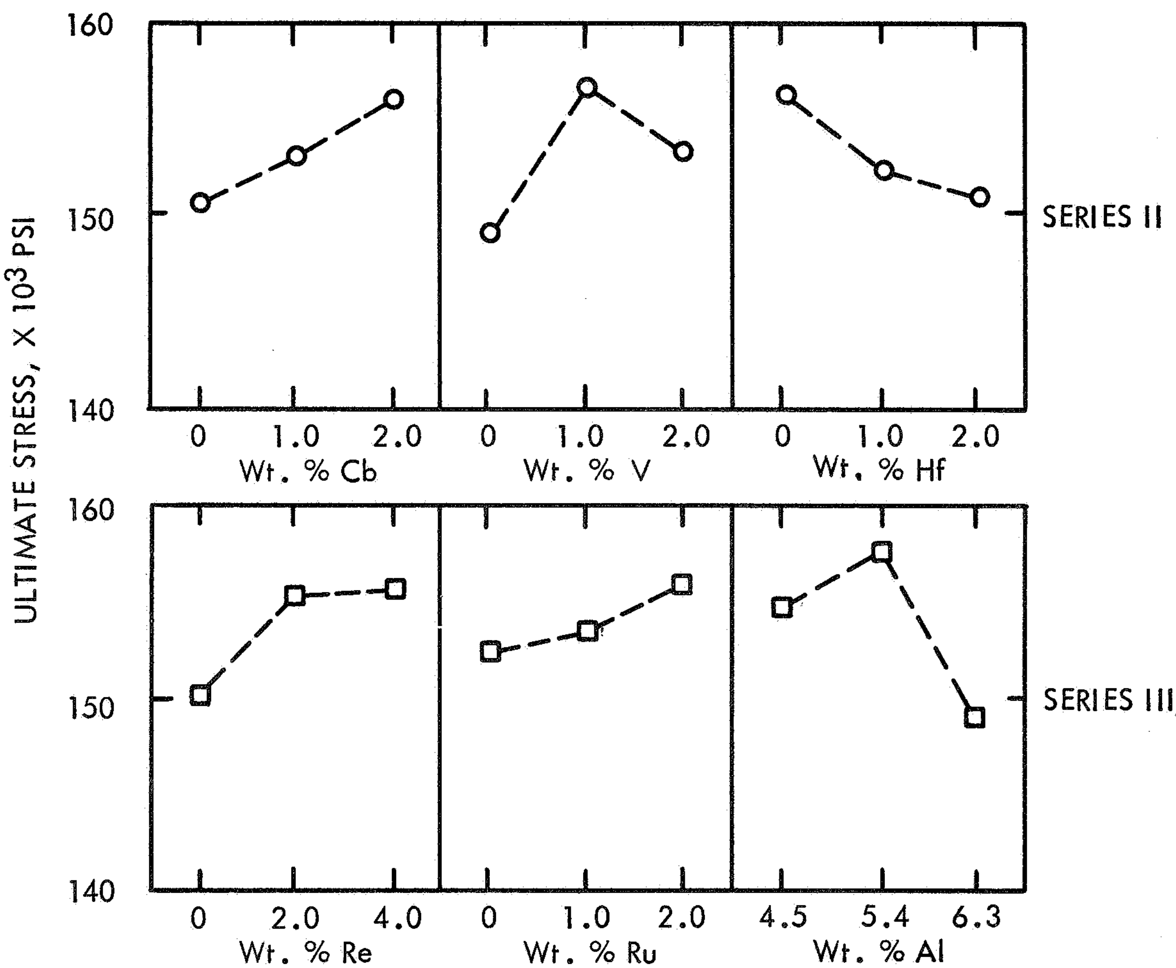


Figure 21. Average Room Temperature Tensile Strength for Nickel-Base Alloys of Series II and III. (Solid Lines are Statistically Significant.)

1974A

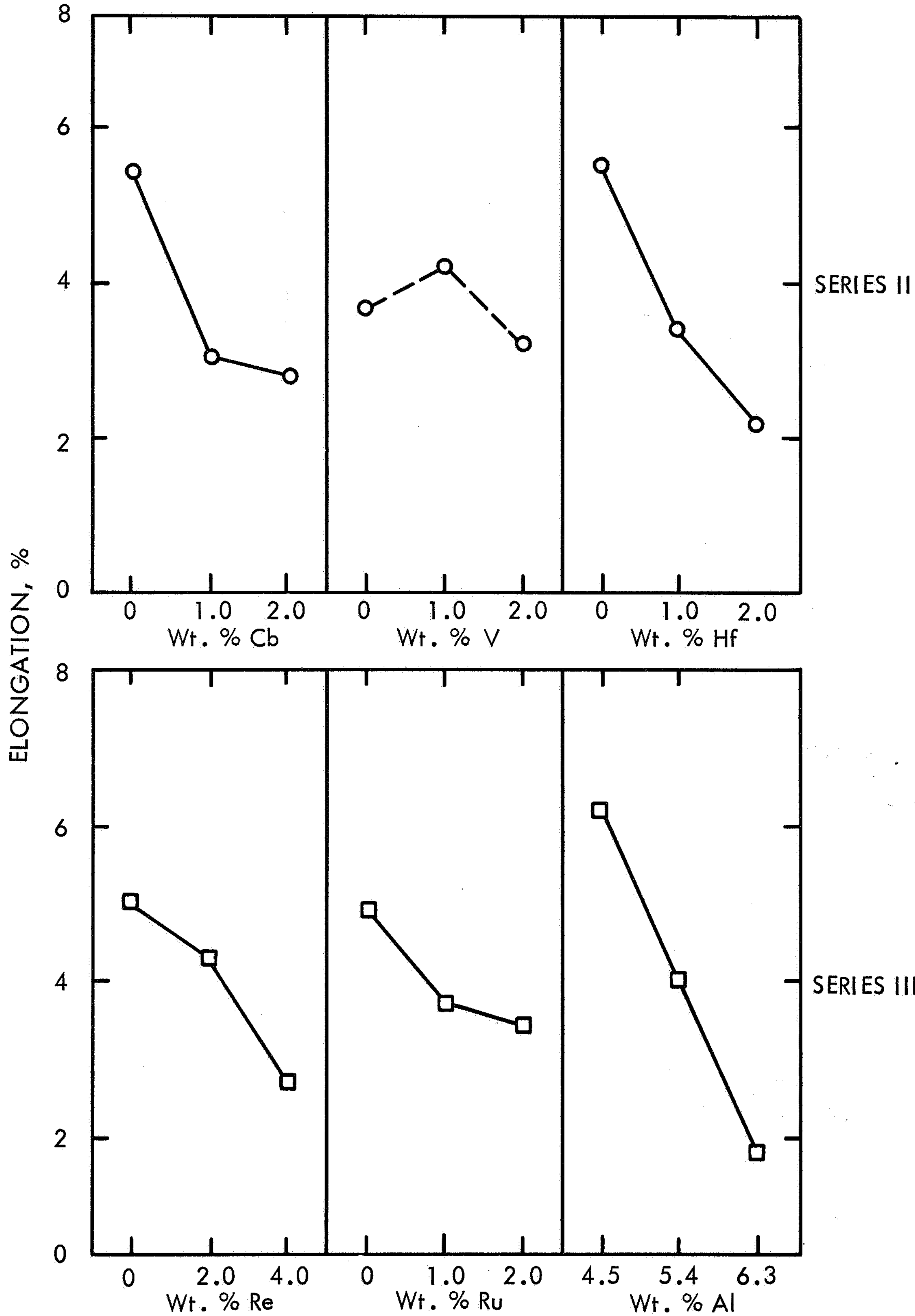


Figure 22. Average Room Temperature Tensile Elongation for Nickel-Base Alloys of Series II and III. (Solid Lines are Statistically Significant.)

1928A

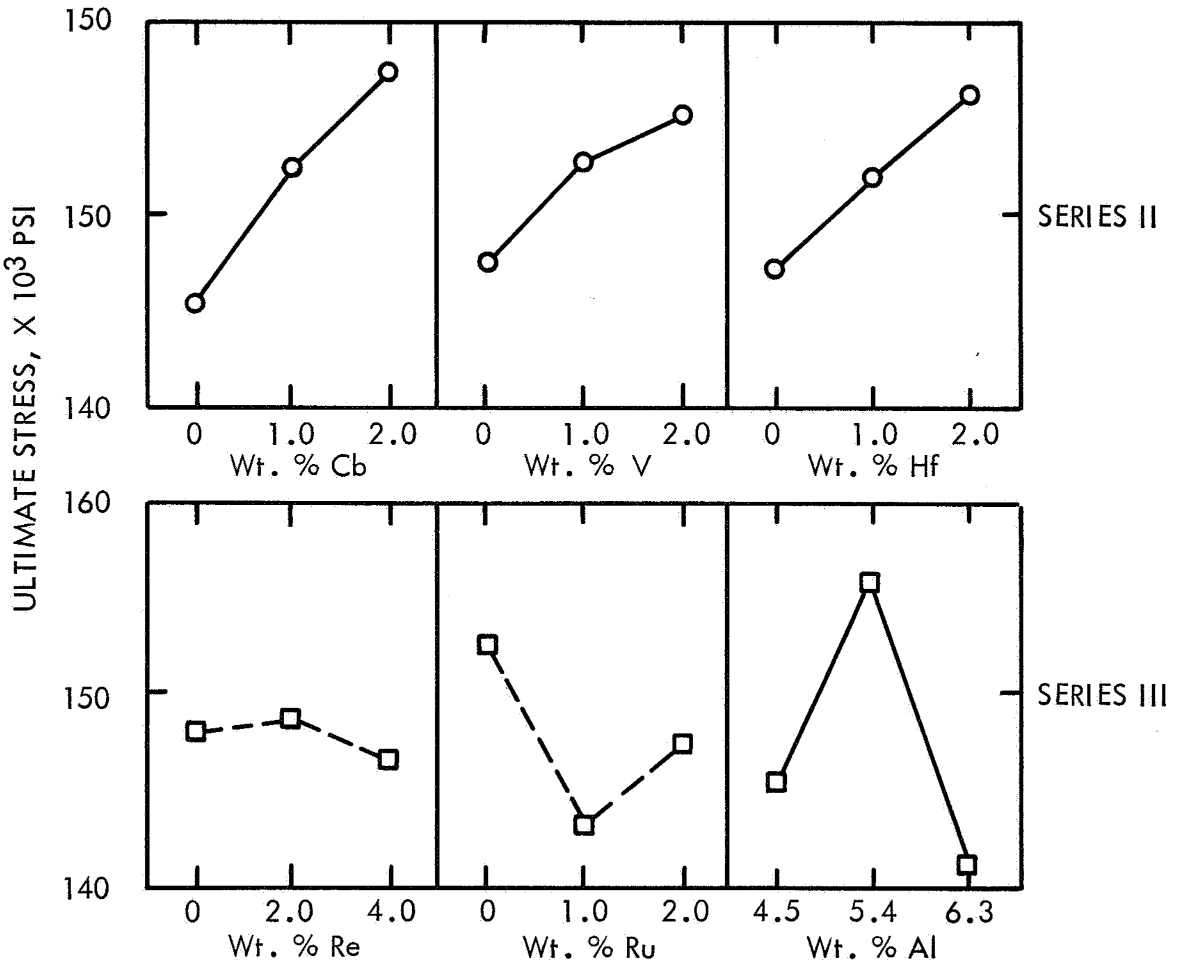


Figure 23. Average 1400°F Tensile Strength for Nickel-Base Alloys of Series II and III. (Solid Lines are Statistically Significant.)

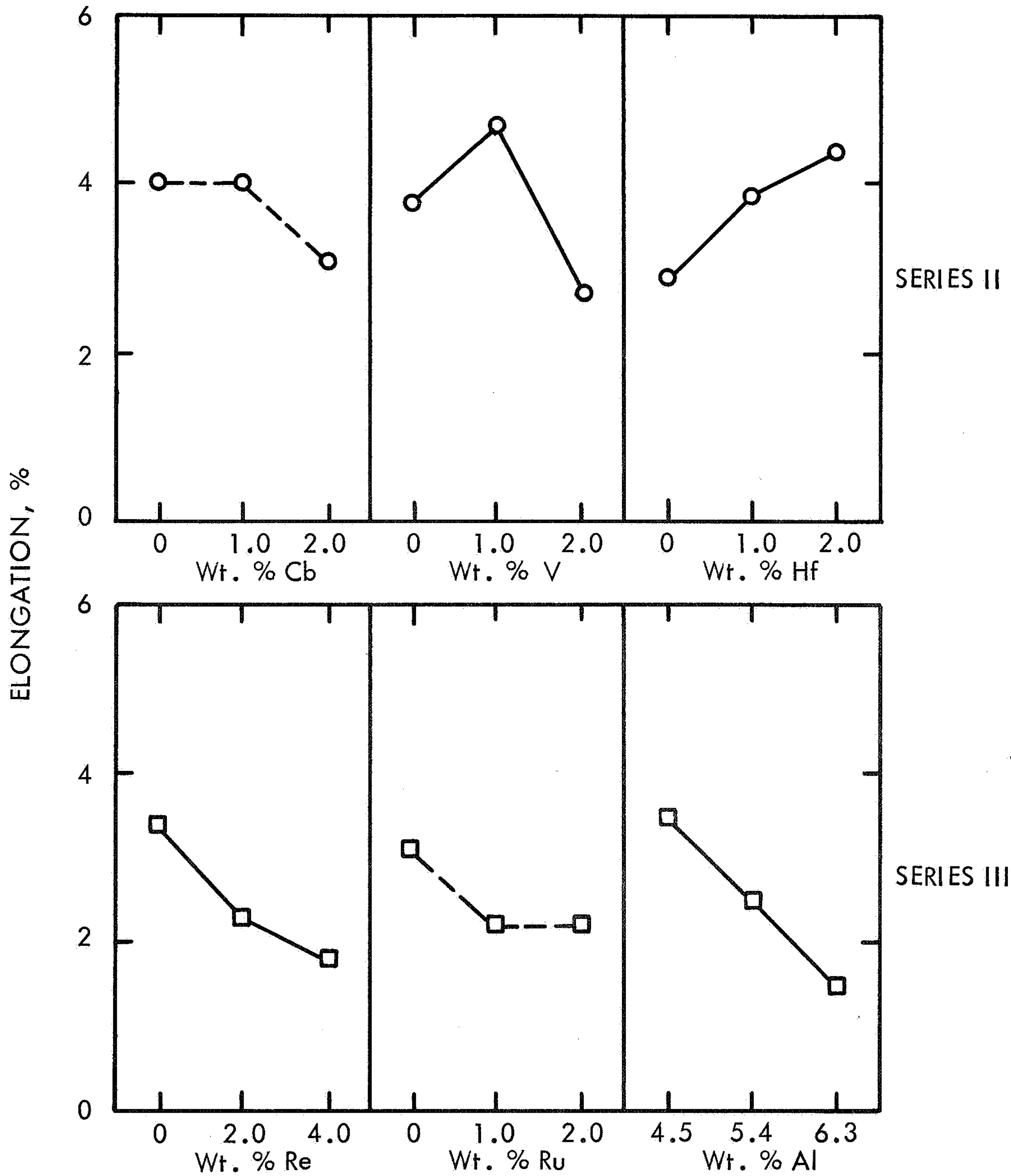


Figure 24. Average 1400°F Elongation for Nickel-Base Alloys of Series II and III. (Solid Lines are Statistically Significant.)

1977A

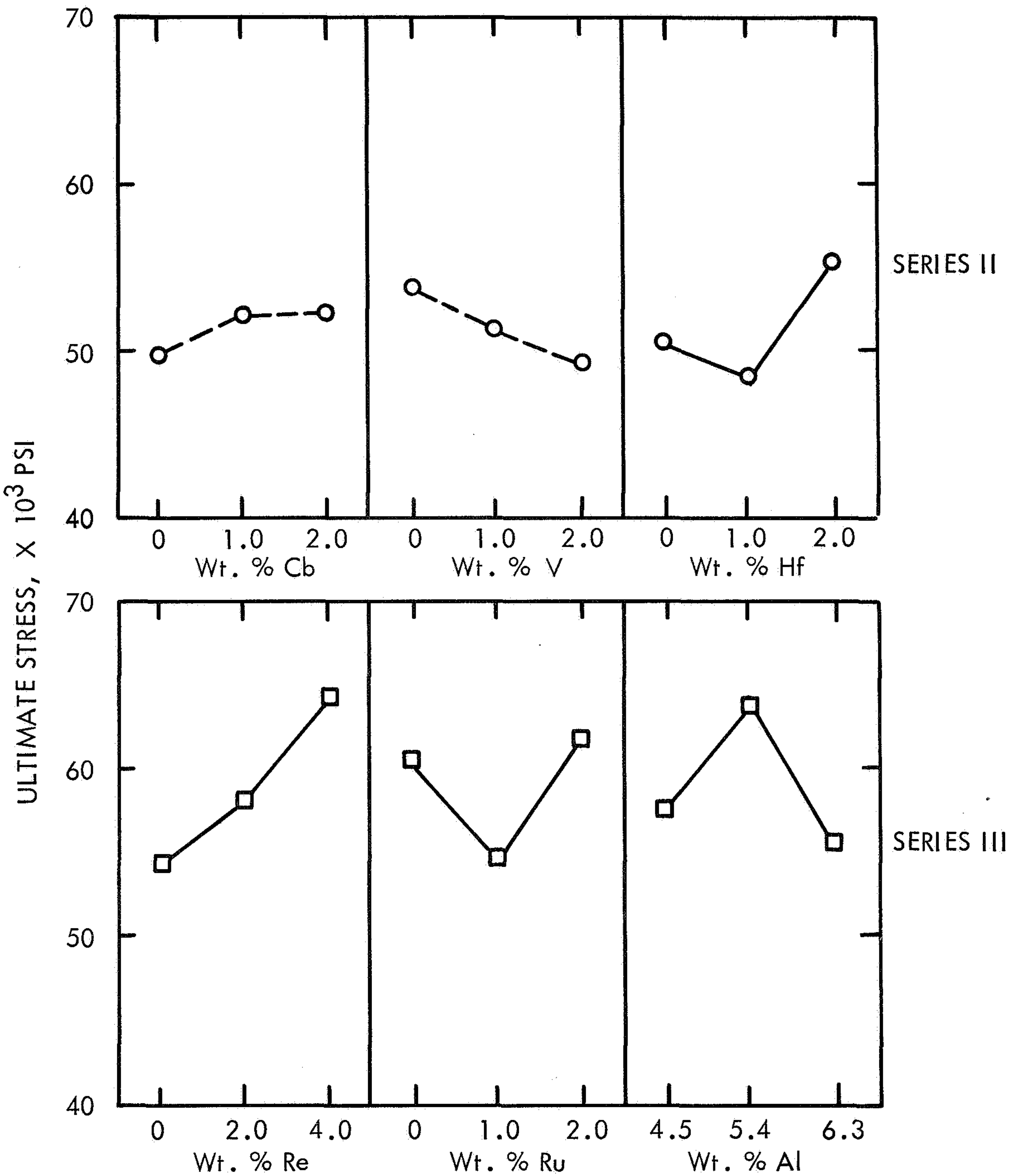


Figure 25. Average 1875°F Tensile Strength for Nickel-Base Alloys of Series II and III. (Solid Lines are Statistically Significant.)

1971A

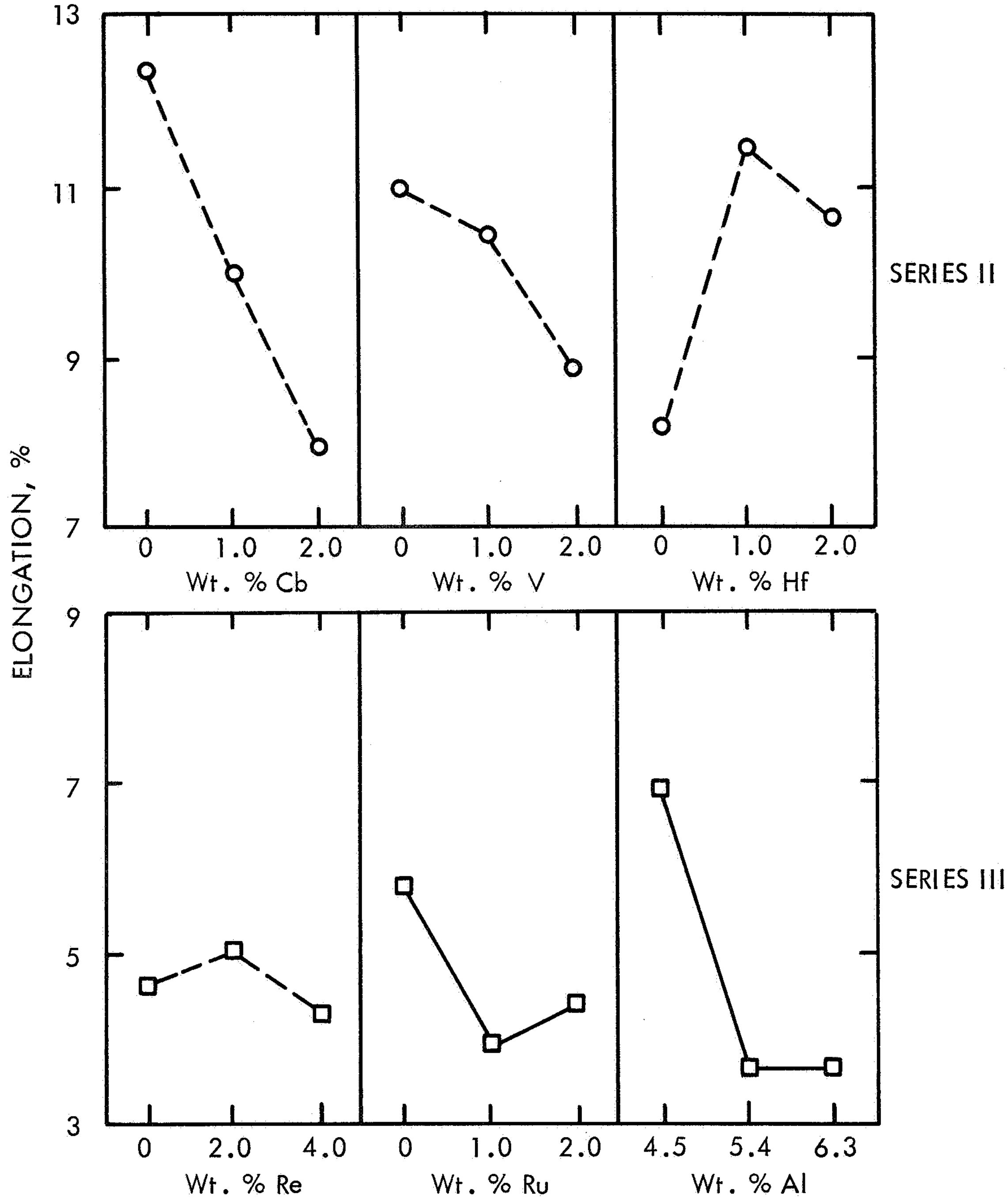


Figure 26. Average 1875°F Tensile Elongation for Nickel-Base Alloys of Series II and III. (Solid Lines are Statistically Significant.)

1969A

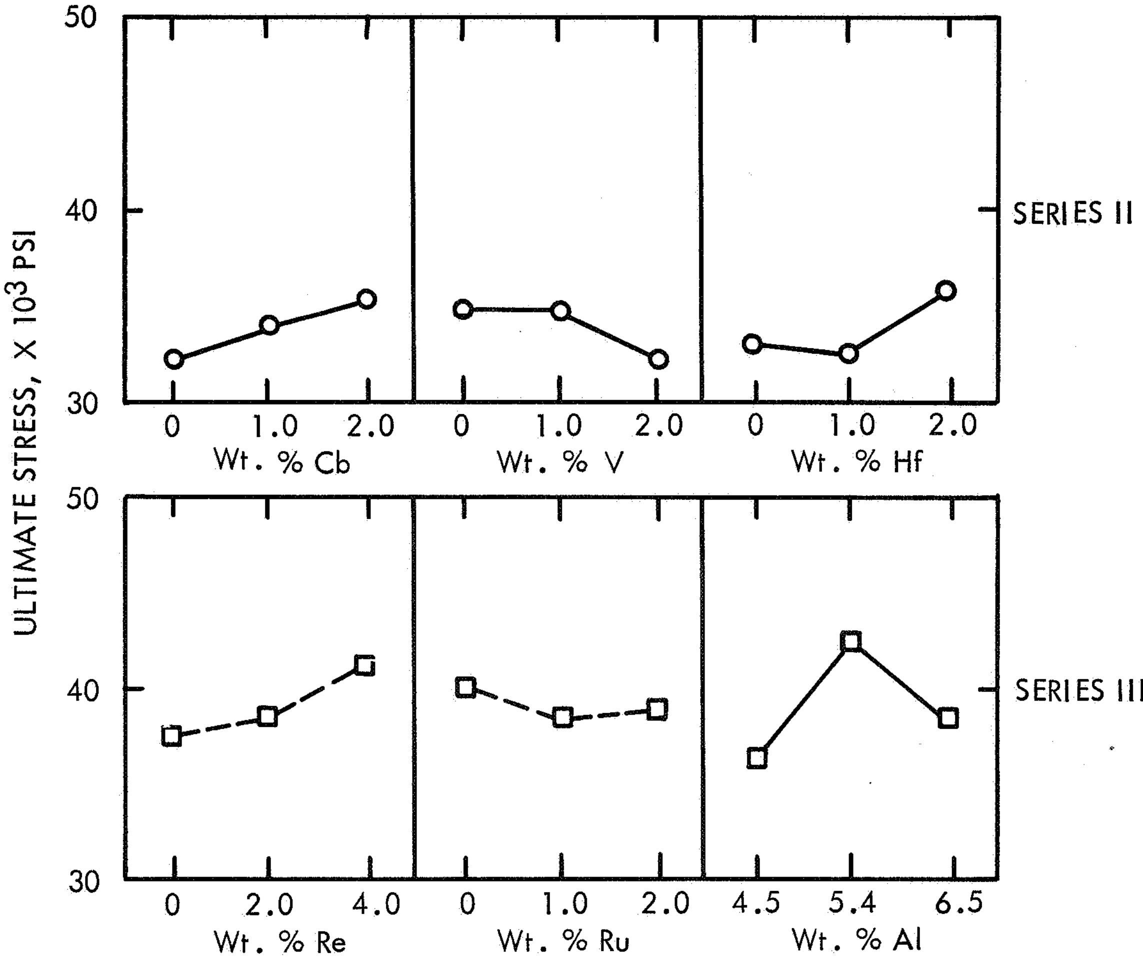


Figure 27. Average 2000°F Tensile Strength for Nickel-Base Alloys of Series II and III. (Solid Lines are Statistically Significant.)

19321A

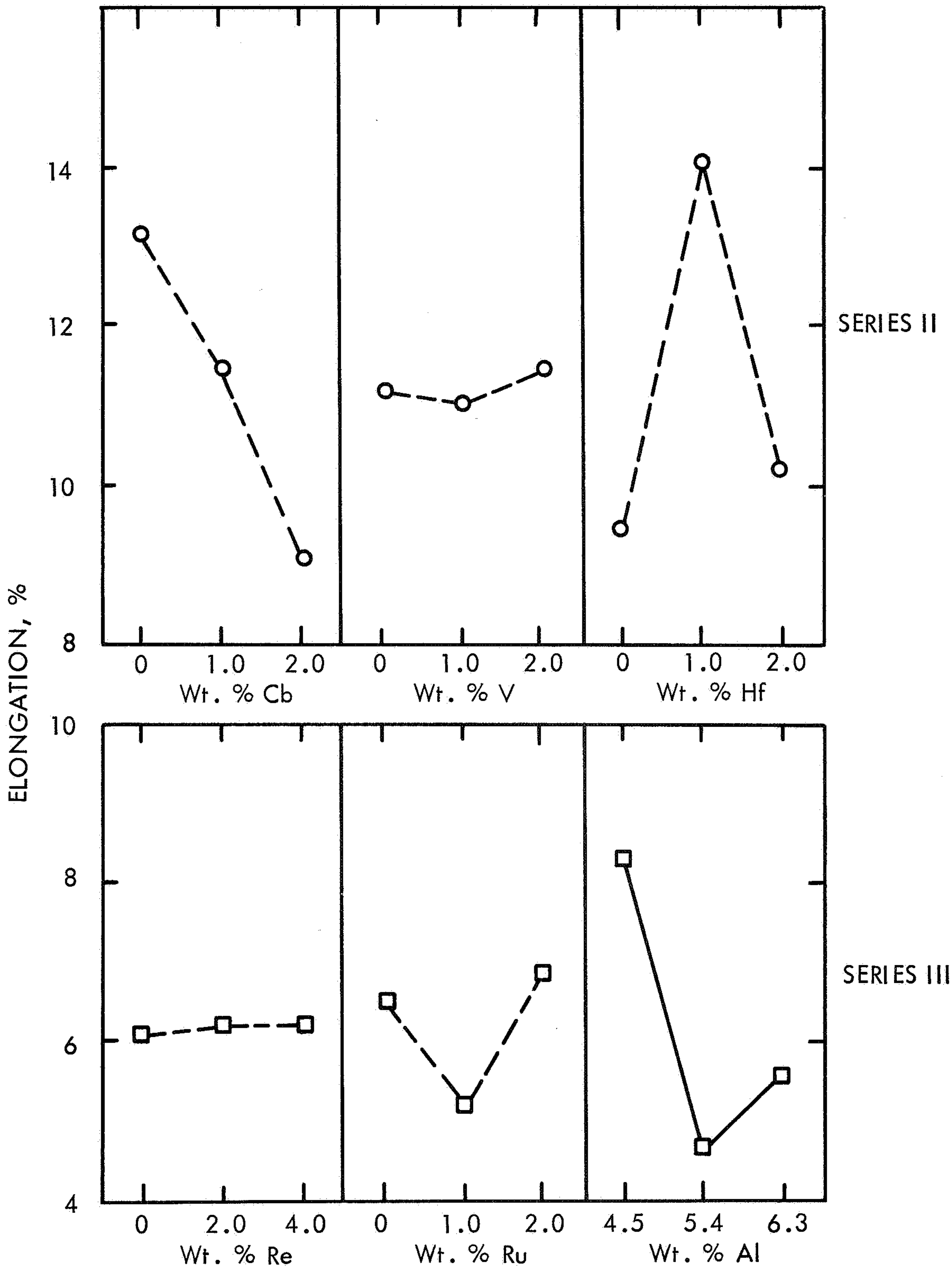
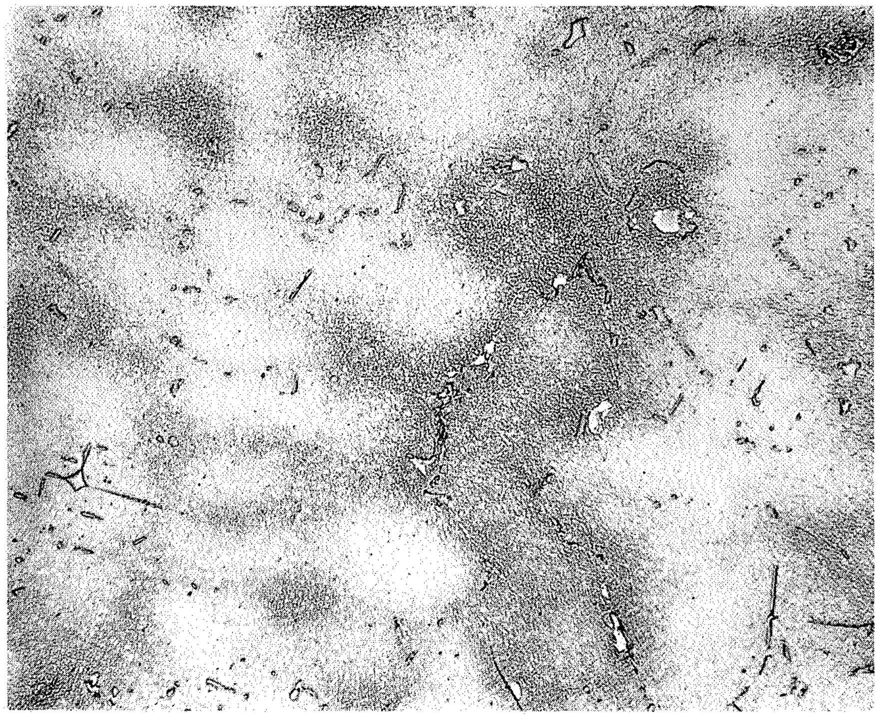
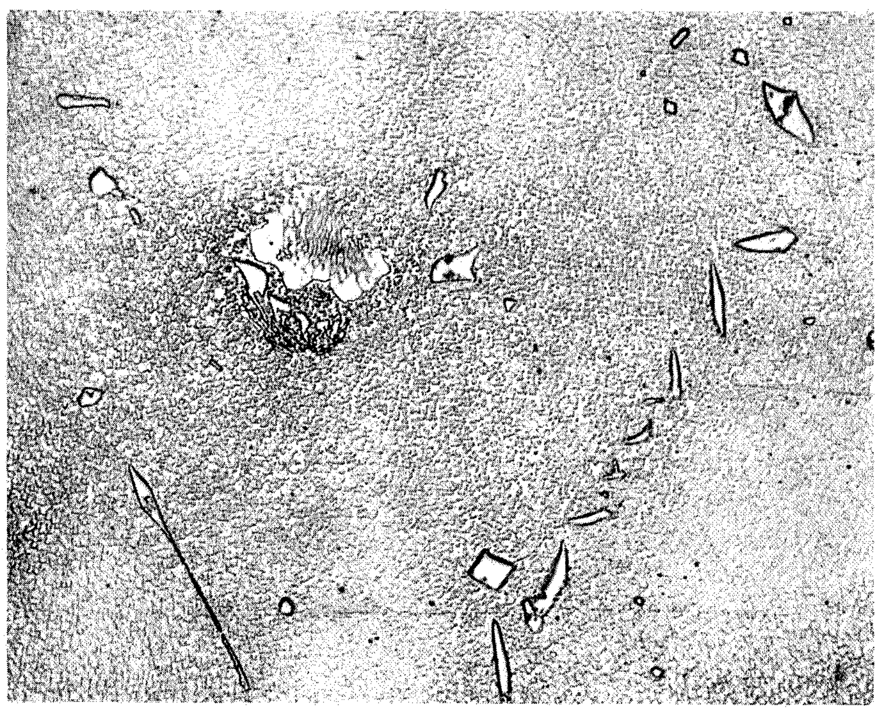


Figure 28. Average 2000°F Tensile Elongation for Nickel-Base Alloys of Series II and III. (Solid Lines are Statistically Significant.)

1930A



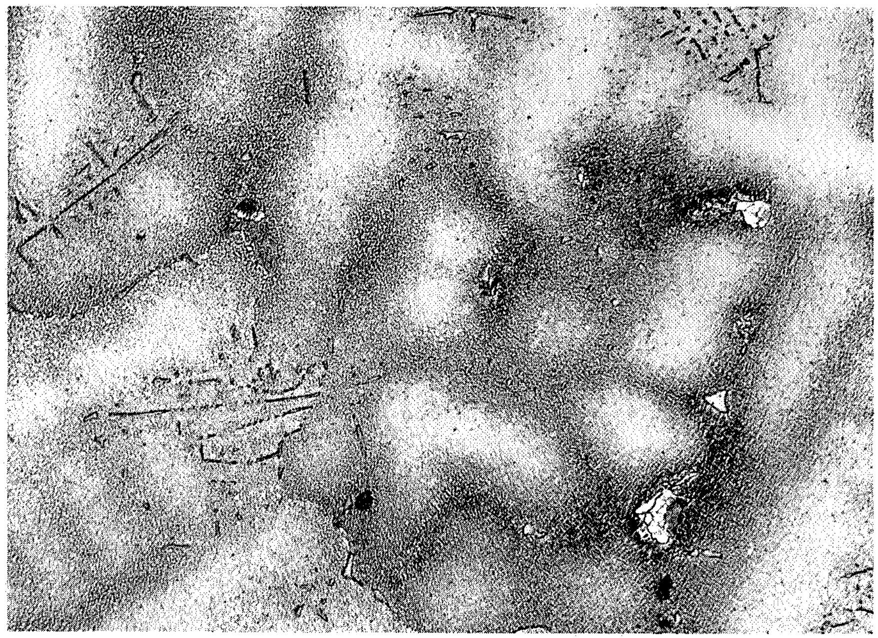
A. 250X



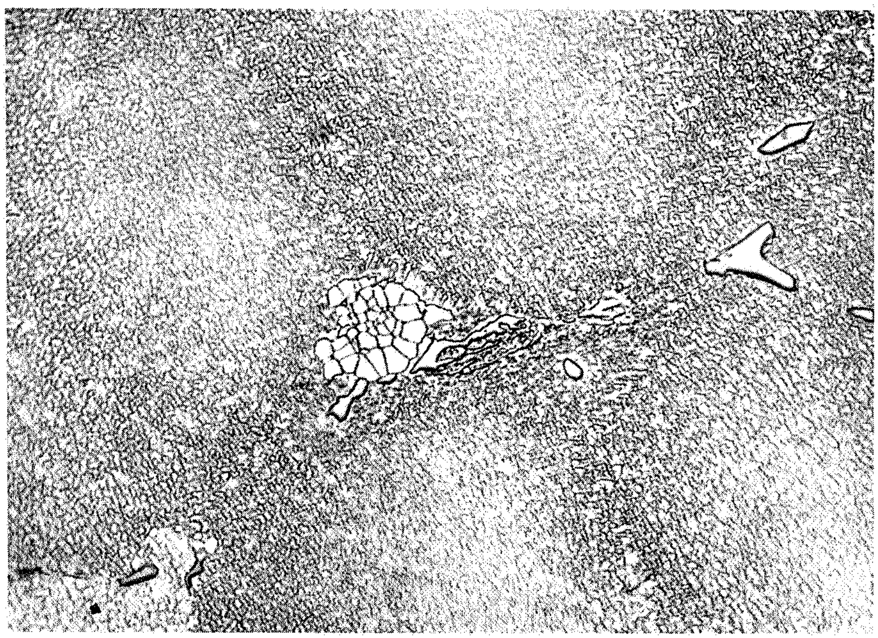
B. 750X

Figure 29. Microstructure of Nickel-Base Alloy II f Showing Carbide and Gamma-Prime Formations. Average 2000°F Stress Rupture Life, 5.0 Hours.
Etchant: 62% H_2O, 15% HF, 15% H_2SO_4, and 8% HNO_3.

1964A



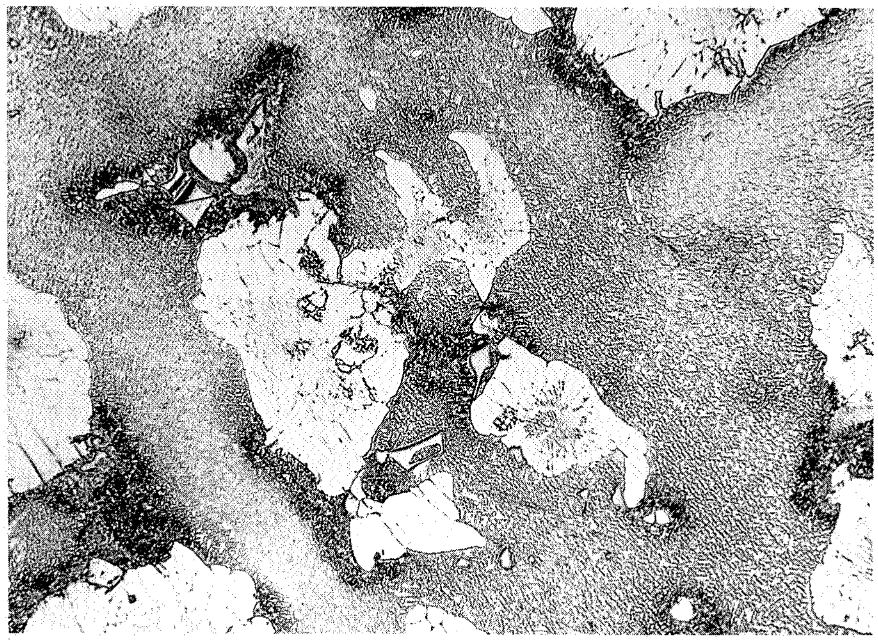
A. 250X



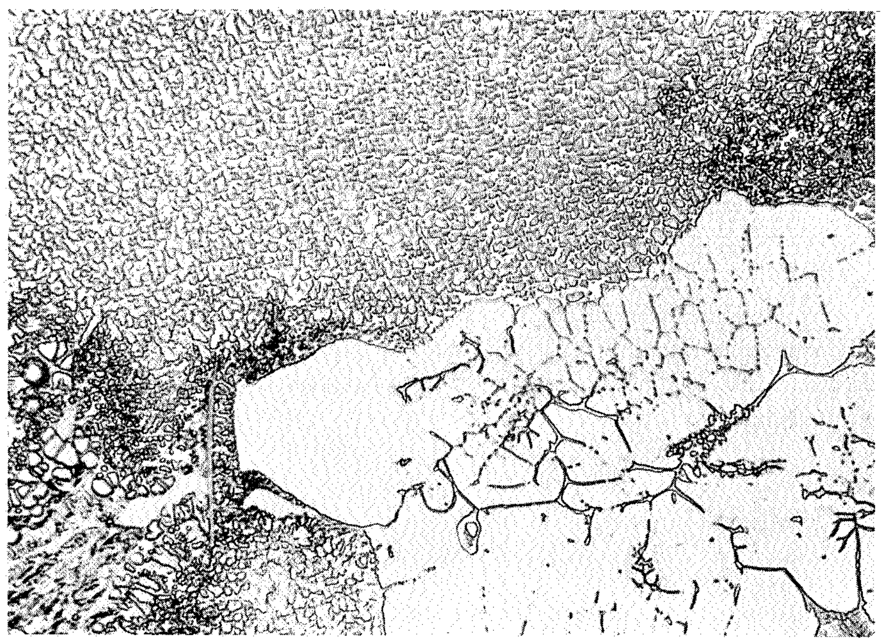
B. 750X

Figure 30. Microstructure of Nickel-Base Alloy III g_o Showing Carbide and Gamma-Prime Formations. Average 2000^{o}F Stress Rupture Life, 15.5 hours.
Etchant: 62% H_2O, 15% HF, 15% H_2SO_4, and 8% HNO_3.

1965A



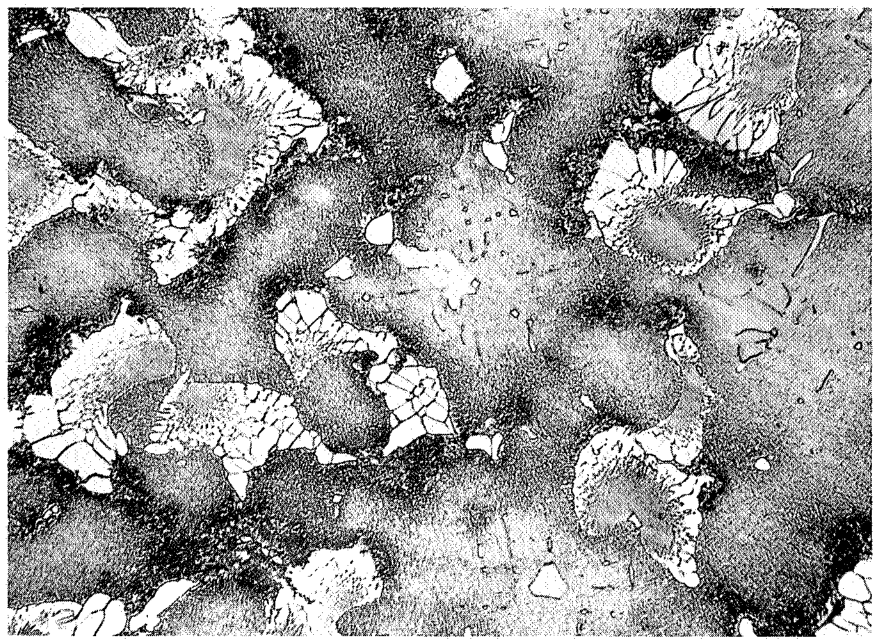
A. 250X



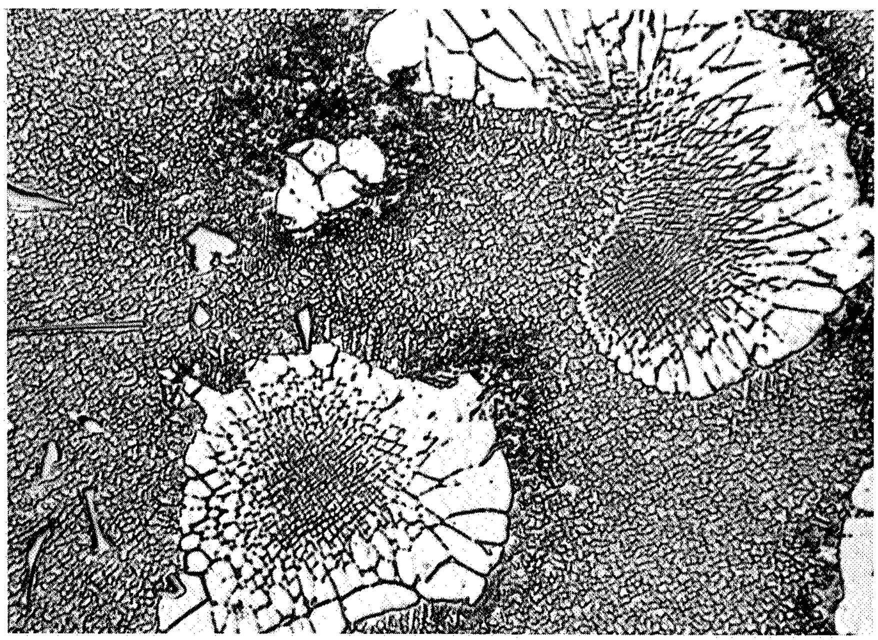
B. 750X

Figure 31. Microstructure of Nickel-Base Alloy IV Y Showing Carbide and Gamma-Prime Formations. Average 2000°F Stress Rupture Life, 41.3 hours.
Etchant: 62% H_2O, 15% HF, 15% H_2SO_4, and 8% HNO_3.

1967A



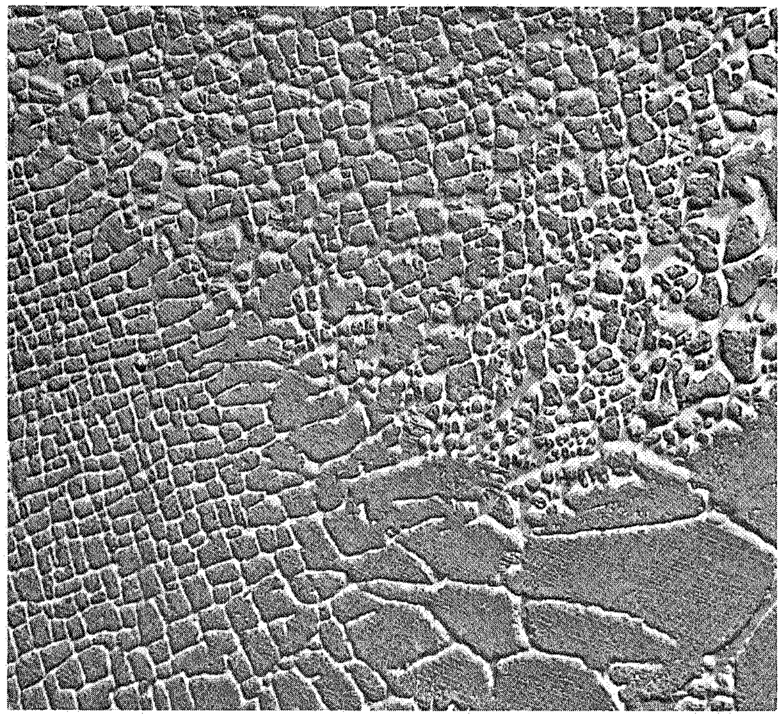
A. 250X



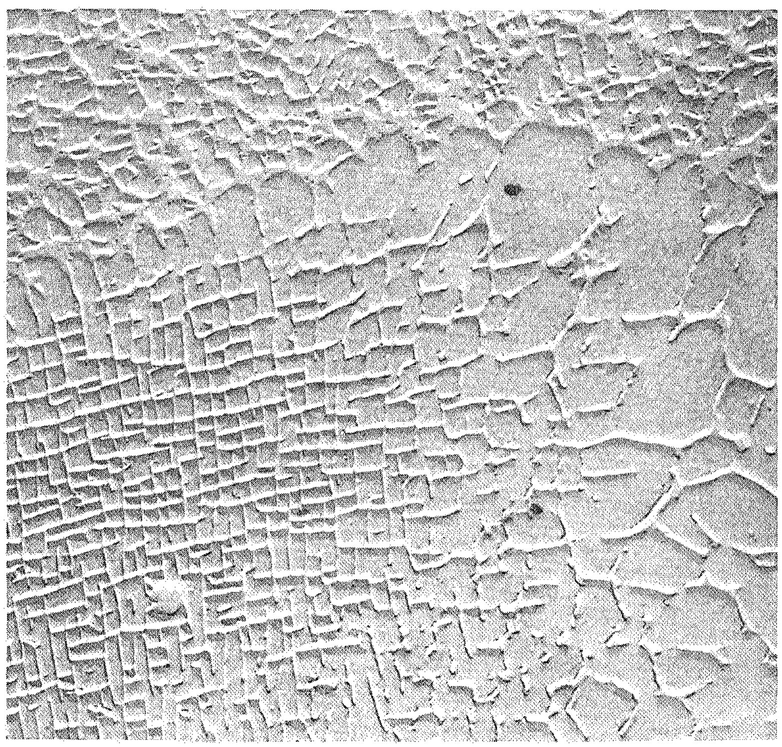
B. 750X

Figure 32. Microstructure of Nickel-Base Alloy VI D Showing Carbide and Gamma-Prime Formations. Average 2000°F Stress Rupture Life 67.3 hours.
Etchant: 62% H_2O, 15% HF, 15% H_2SO_4, and 8% HNO_3.

1963A



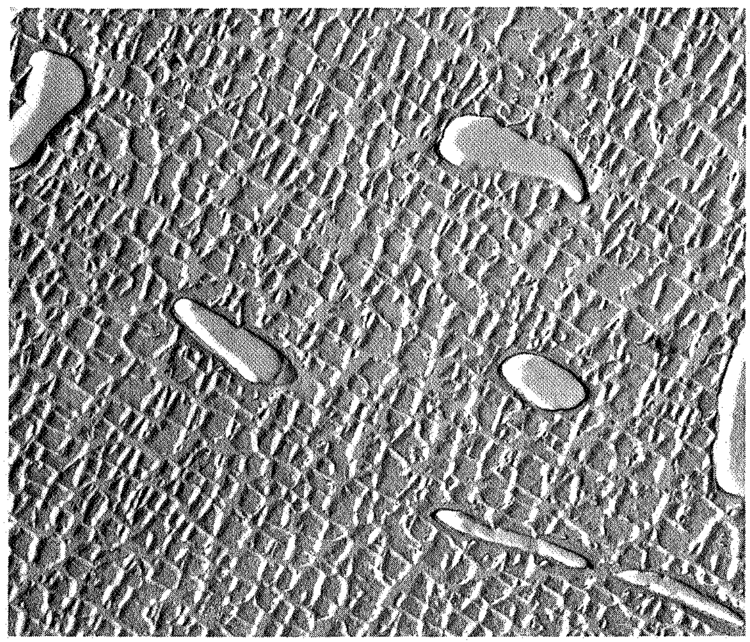
A. Alloy V G



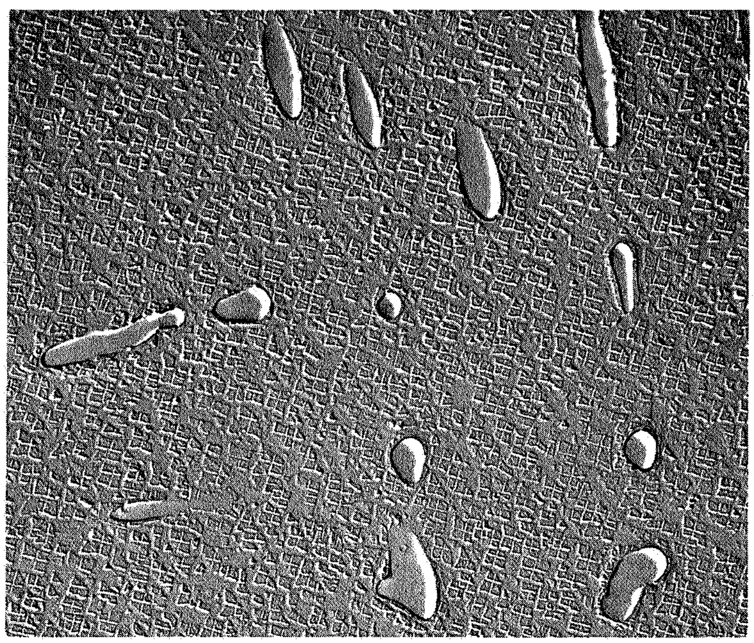
B. Alloy VI D

Figure 33. Typical Massive Gamma-Prime Regions of Nickel-Base Alloys with Superior Stress Rupture Properties. 4,000X.

1966A



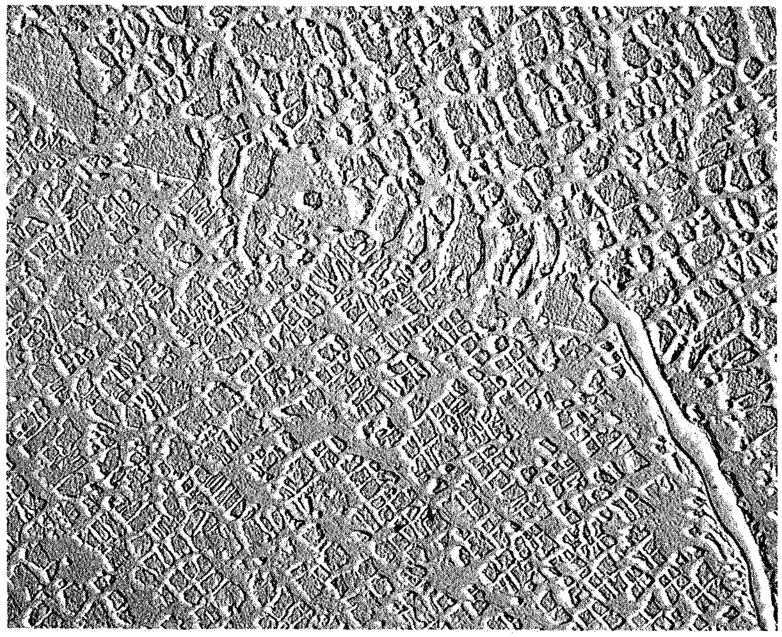
A. Alloy VI D. Massive MC Type Carbides



B. Alloy III g. Massive MC Type Carbides

Figure 34. Typical Carbide and Secondary Gamma-Prime Regions of Nickel-Base Alloys. 4,000X.

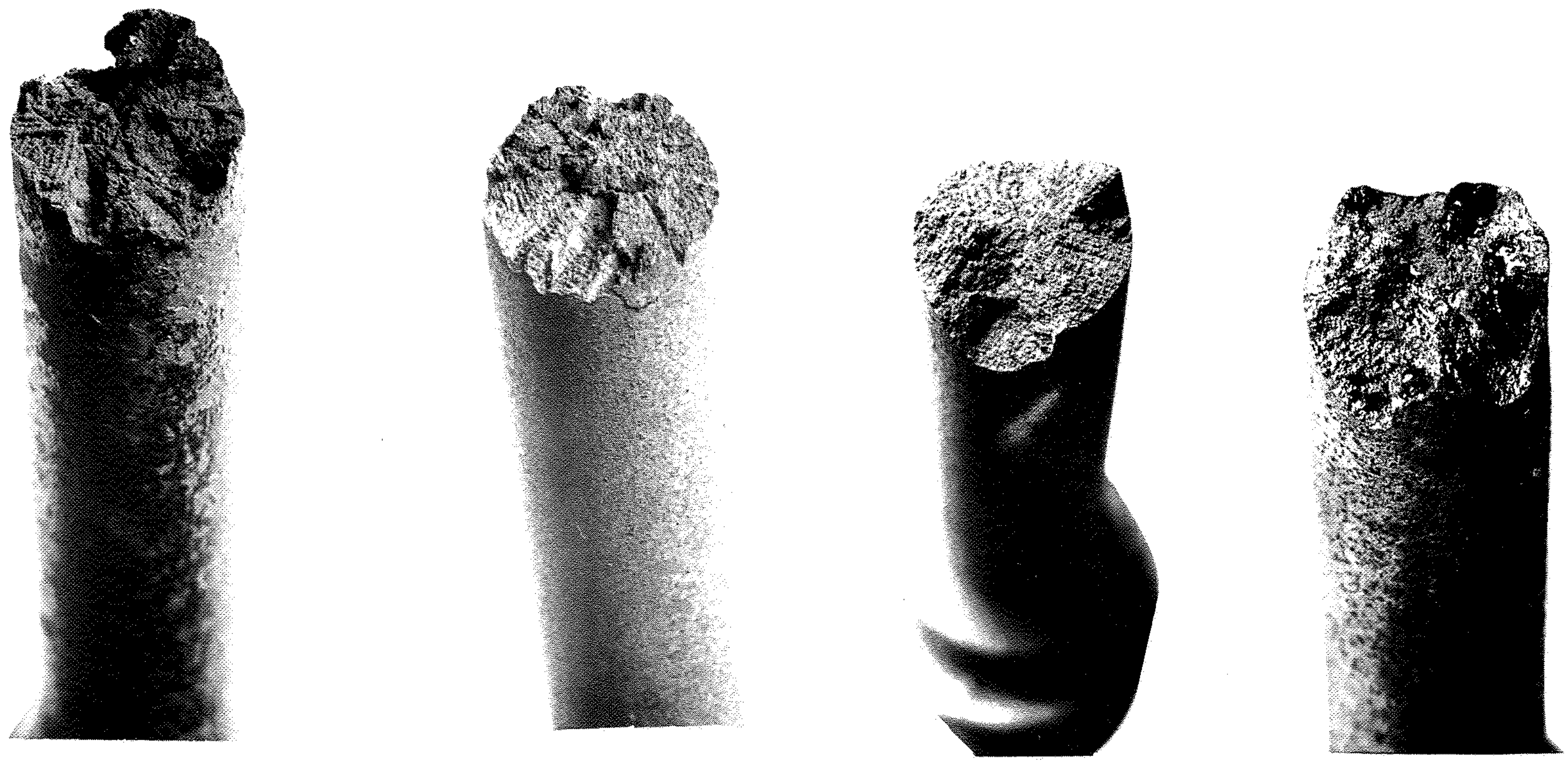
1962A



C. Alloy III g. Possible Grain Boundary Carbide of $M_{23}C_6$ or M_6C Type

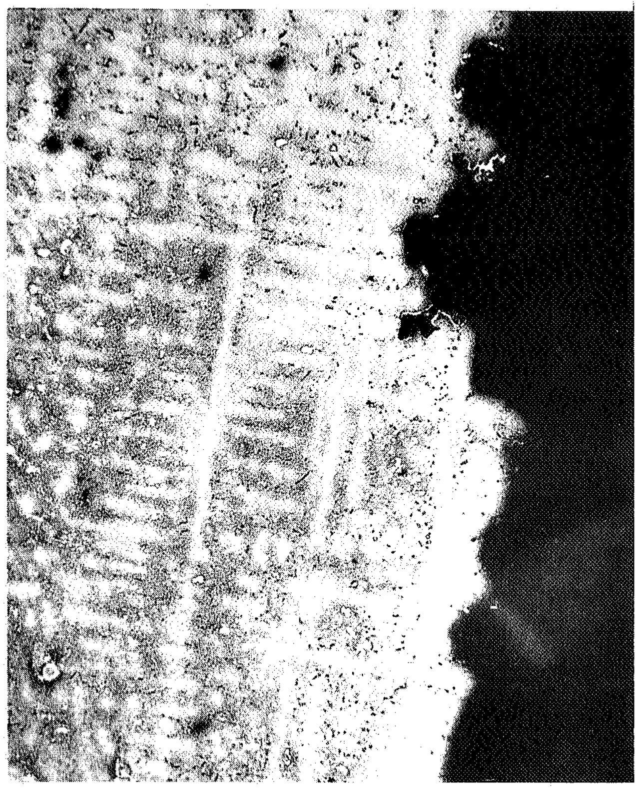
Figure 34. (continued).

1959A



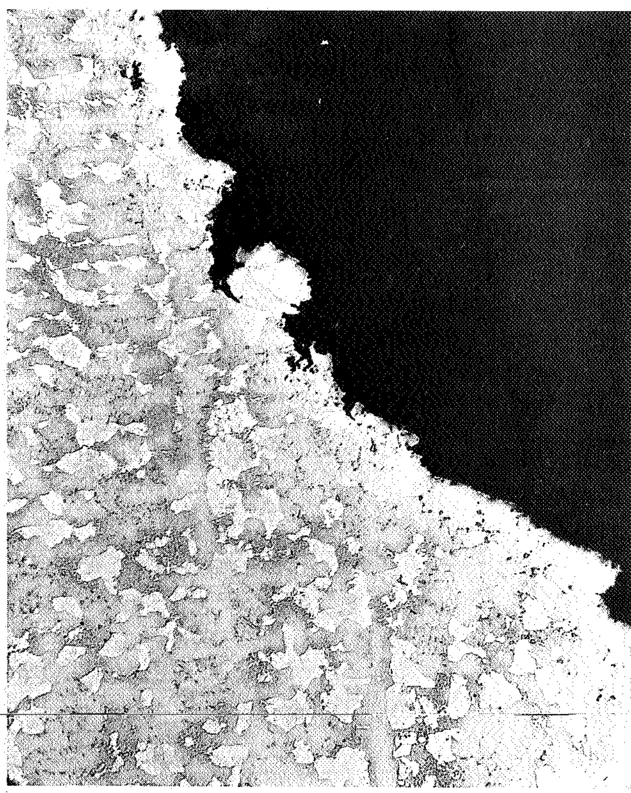
A. Alloy II f   B. Alloy III g   C. Alloy IV Y   D. Alloy VI D

Figure 35. Appearance of Typical 2000°F Stress Rupture Fractures in Experimental Nickel-Base Alloys. 5X.

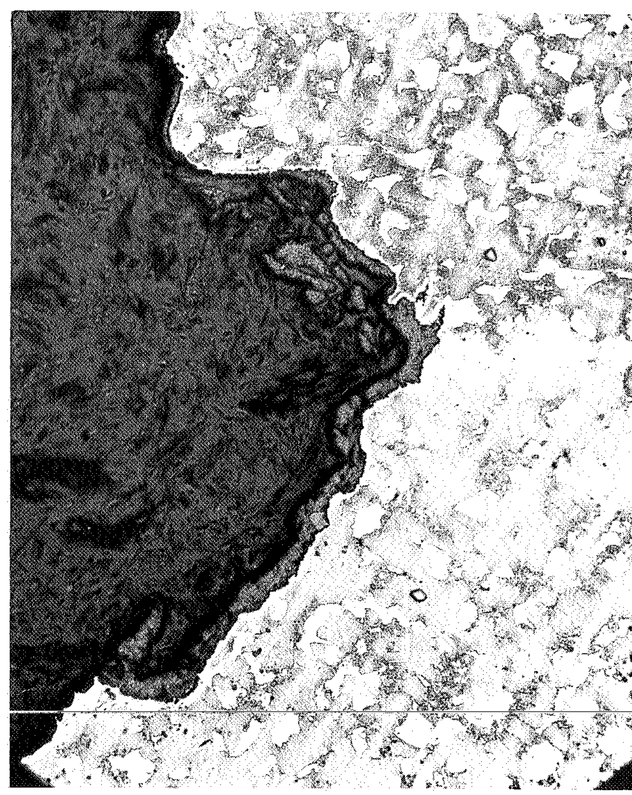


A. Alloy II f

B. Alloy III g



C. Alloy IV Y



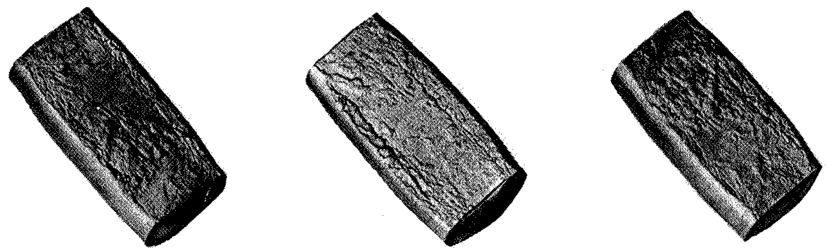
D. Alloy VI D

Figure 36. Microstructural Appearance of Various Nickel-Base Superalloys Adjacent to Fracture Showing Intergranular Nature of 2000°F Stress Rupture Failure. 60X.
Etchant: 62% H_2O, 15% HF, 15% H_2SO_4, and 8% HNO_3.

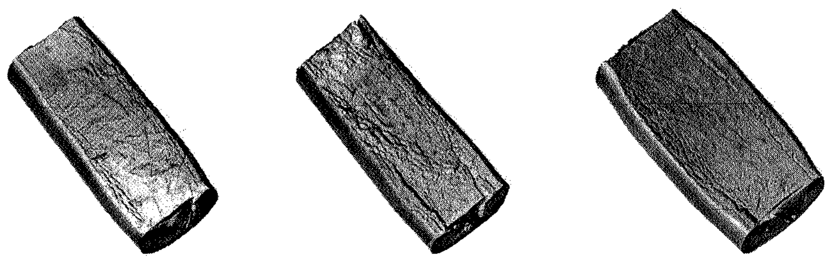
1970A

Figure 37. Typical Extrusions (4:1 Area Reduction) of Experimental Nickel-Base Alloys.

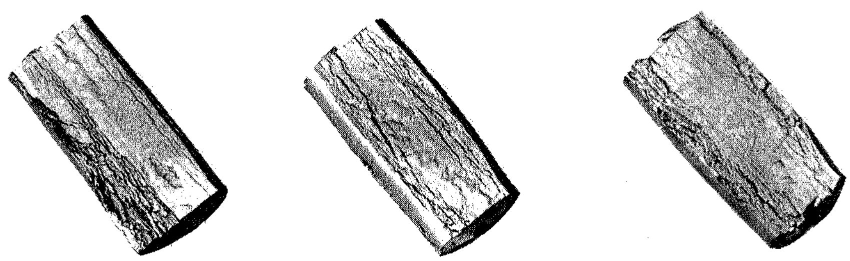
1960A



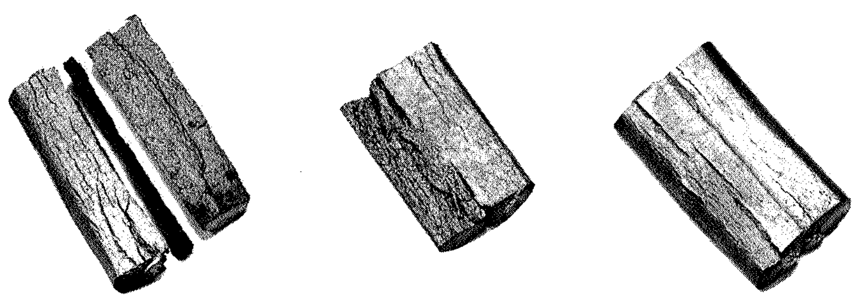
A. Alloy I - III, Rating 2.0



B. Alloy I - XI, Rating 3.0



C. Alloy I - XX, Rating 4.0



D. Alloy I - XXVII, Rating 5.0

Figure 38. **Typical Examples of Sidepressed Bars of Experimental Nickel-Base Alloys for the Various Workability Ratings.** 50% reduction, **Temperatures of Reduction from Left to Right:** 2050°F, 2100°F, and 2150°F.

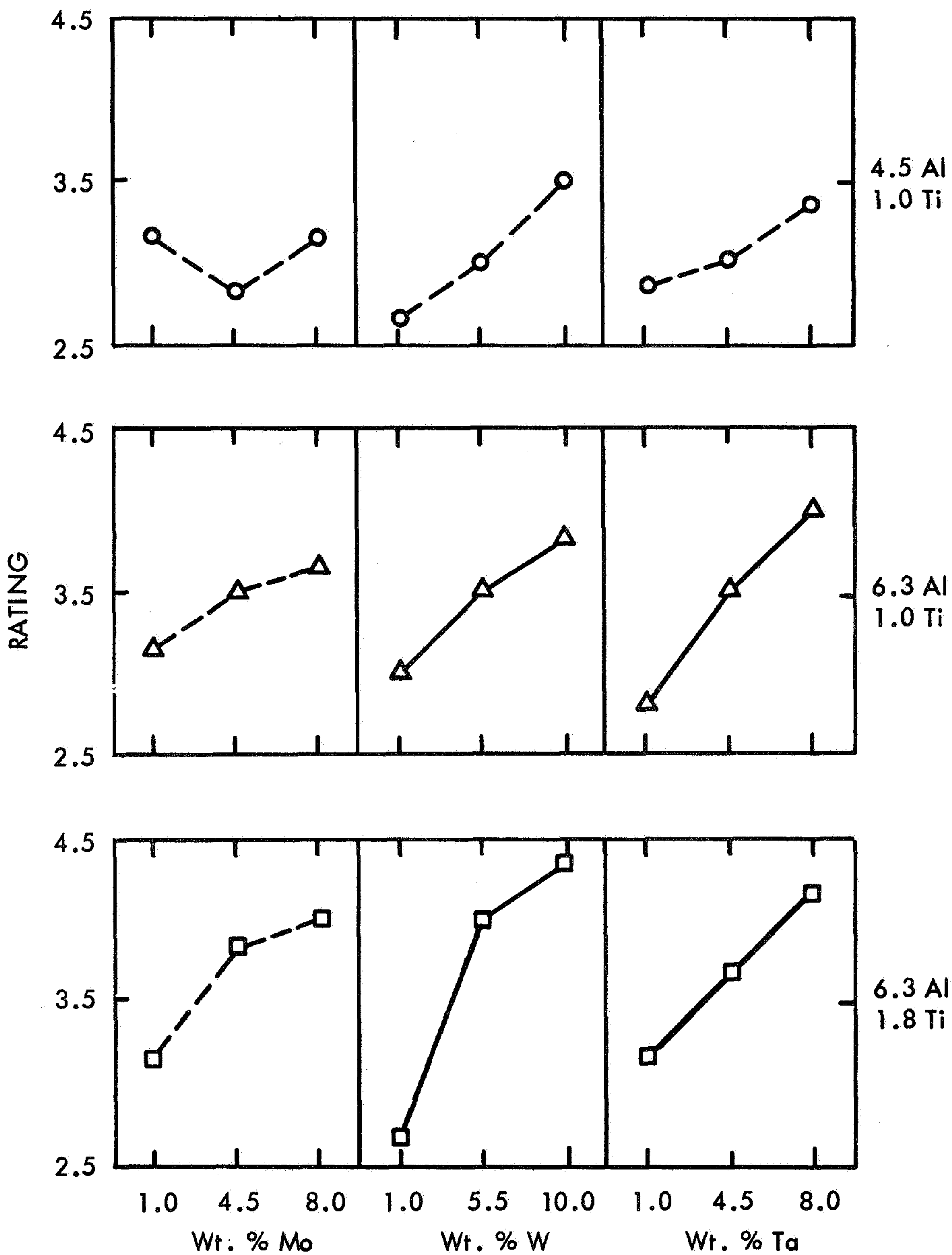


Figure 39. Average Workability Rating for Nickel-Base Alloys of Series I. (Solid Lines are Statistically Significant.)

1973A

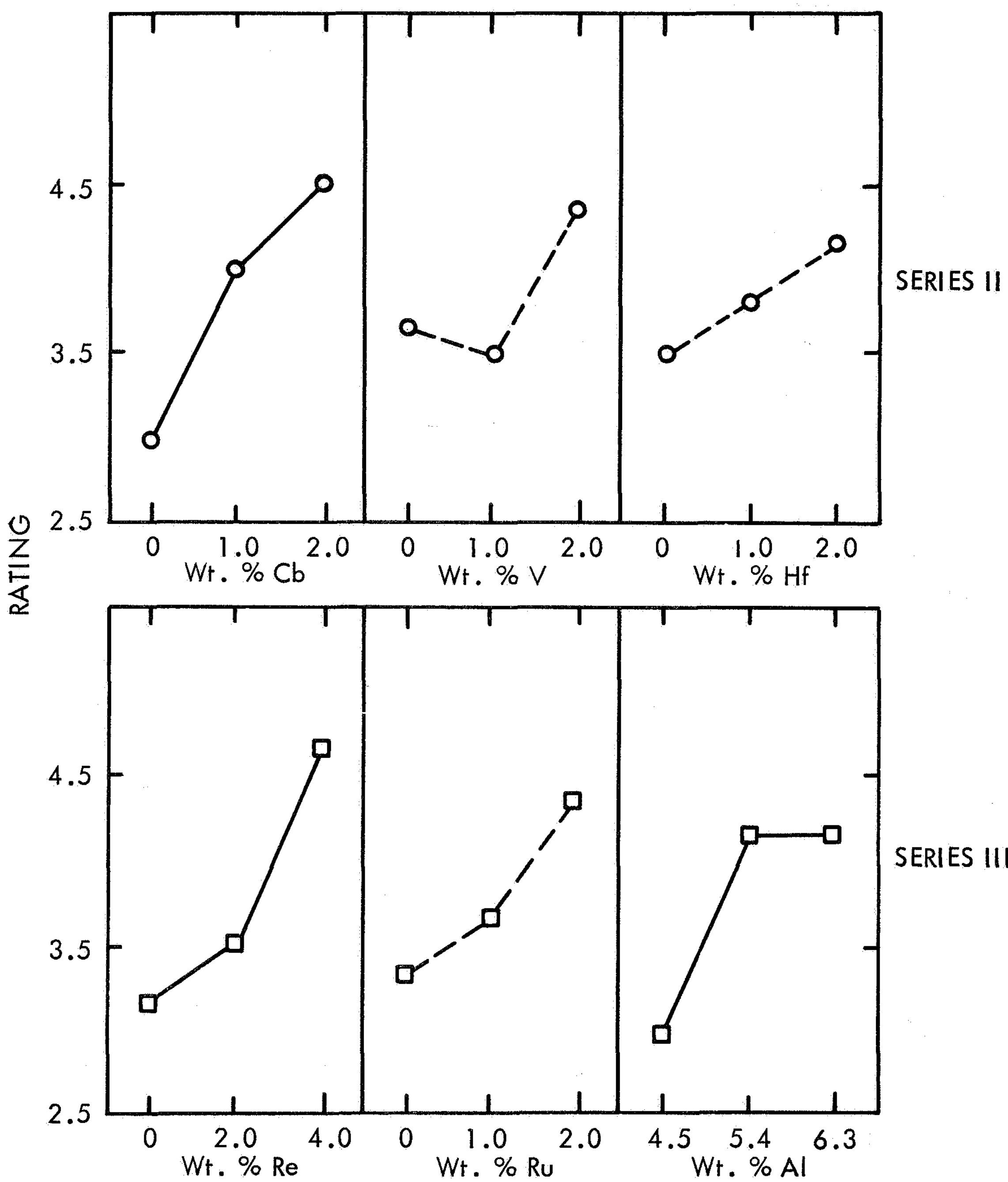


Figure 40. Average Workability Rating for Nickel-Base Alloys of Series II and III. (Solid Lines are Statistically Significant.)

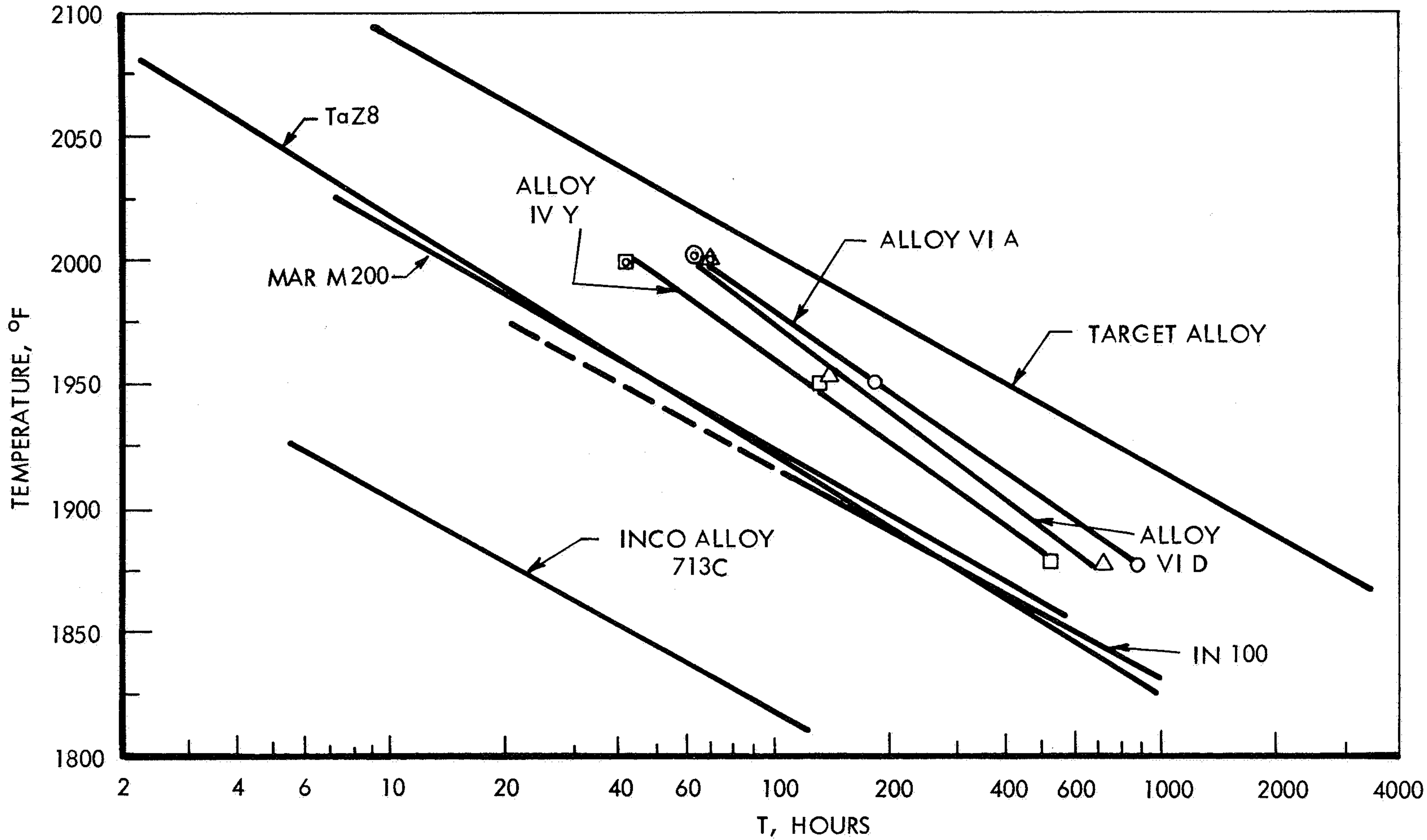


Figure 41. Stress Rupture Properties at 15,000 psi for the Three Cast Experimental Alloys of Task II in Relation to Existent Cast Nickel-Base Superalloys and Target Alloy. (After Freche and Waters[3]).

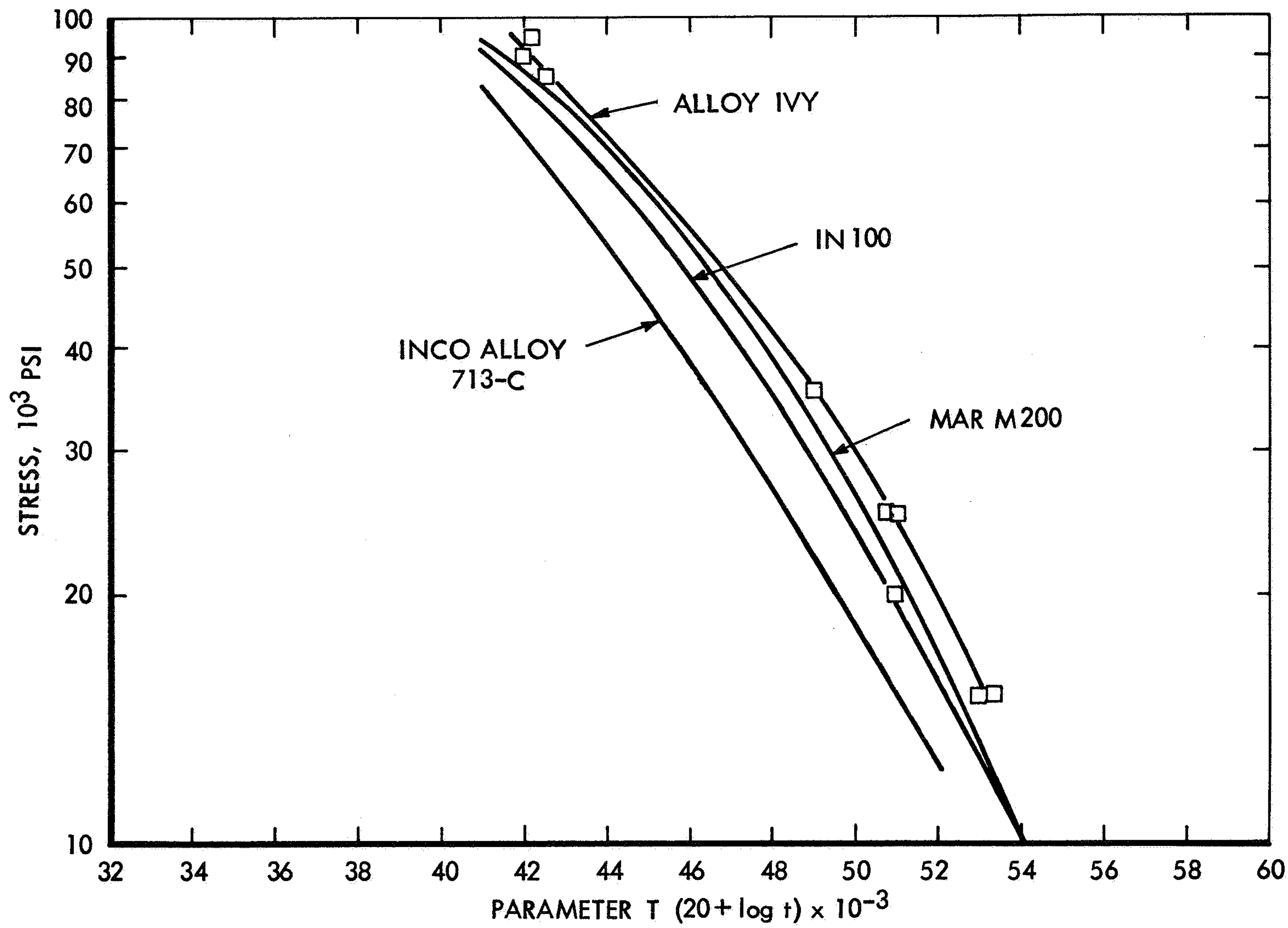


Figure 42. Larson Miller Plot Comparing Alloy IV Y with Existent Cast Nickel-Base Superalloys.

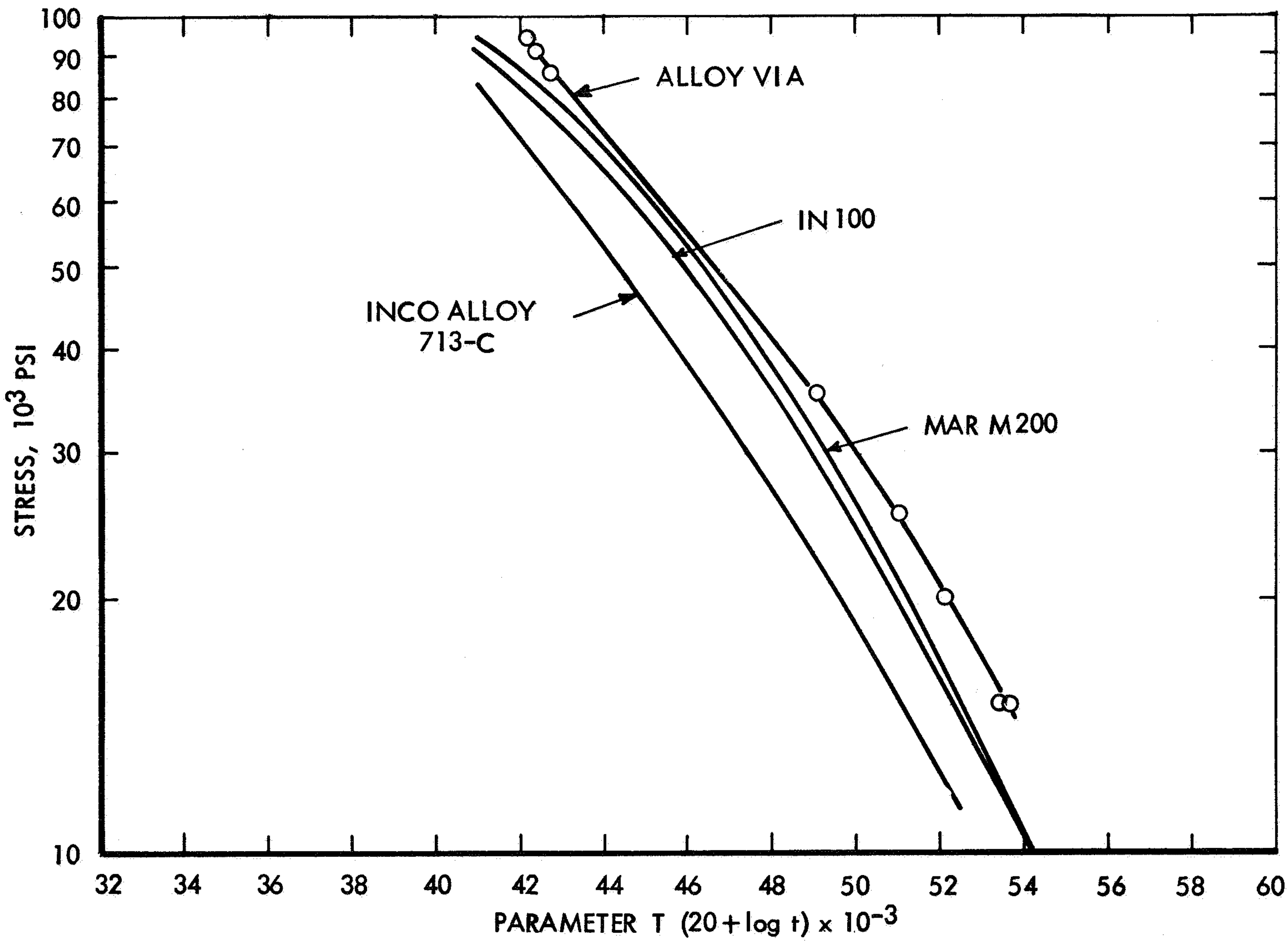


Figure 43. Larson Miller Plot Comparing Alloy VI A with Existent Cast Nickel-Base Superalloys.

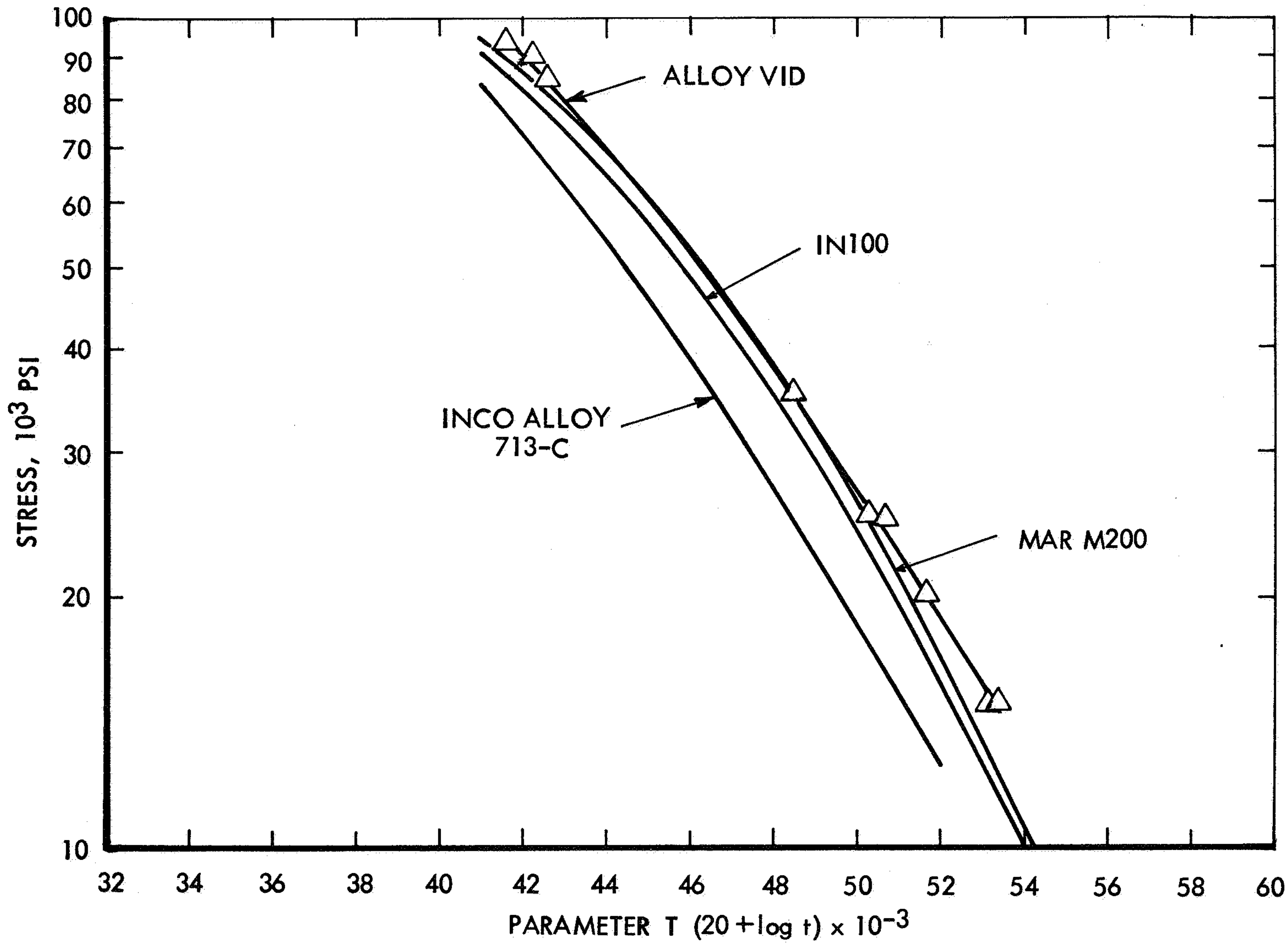


Figure 44. Larson Miller Plot Comparing Alloy VI D with Existent Cast Nickel-Base Superalloys.

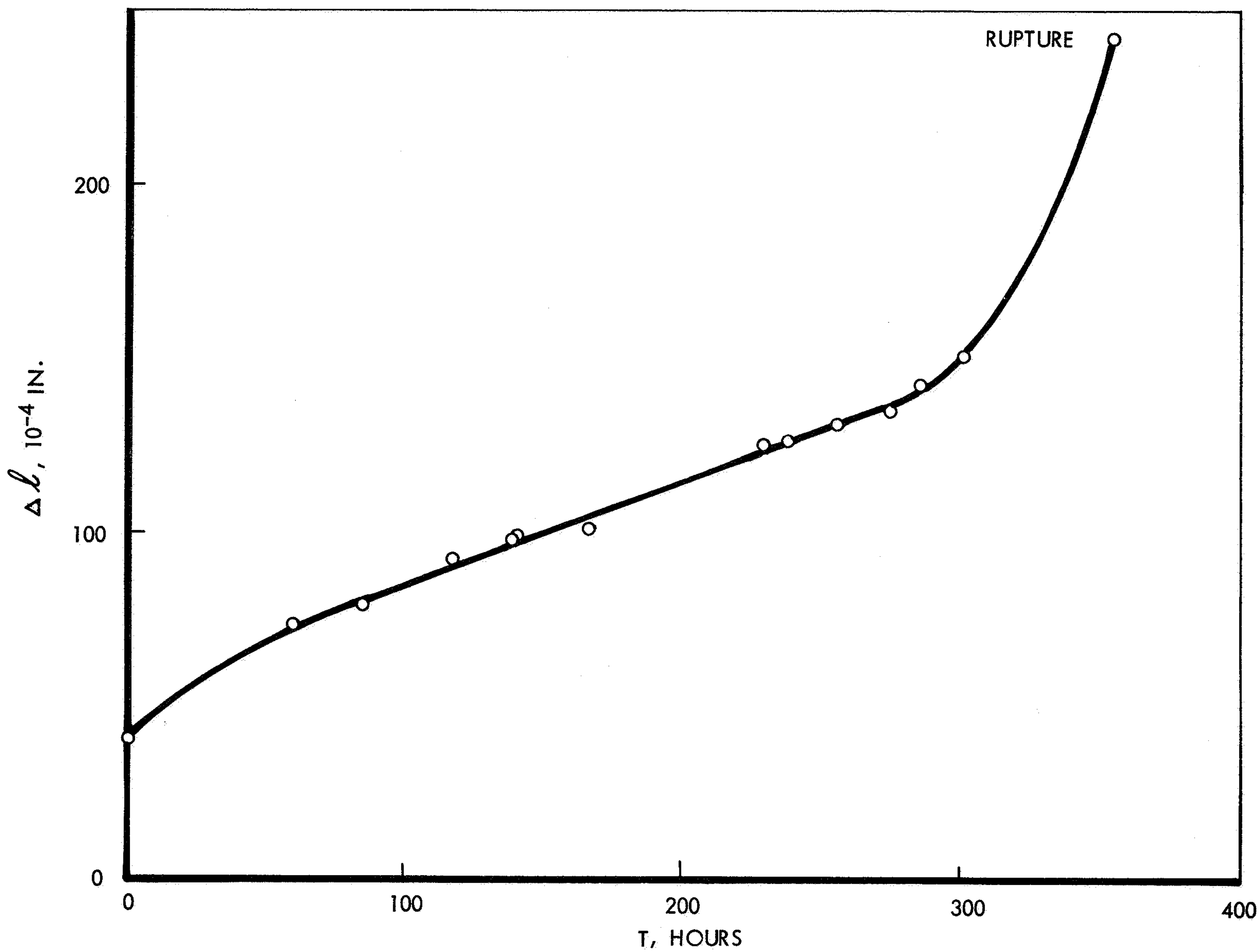


Figure 45. Creep Curve for Alloy IV Y at 1400°F Under 85,000 psi Load Using a Platinum Extensometer. Stress Rupture Life = 355.3 Hours and Percent Elongation = 1.6. Gauge Length: Radius to Radius

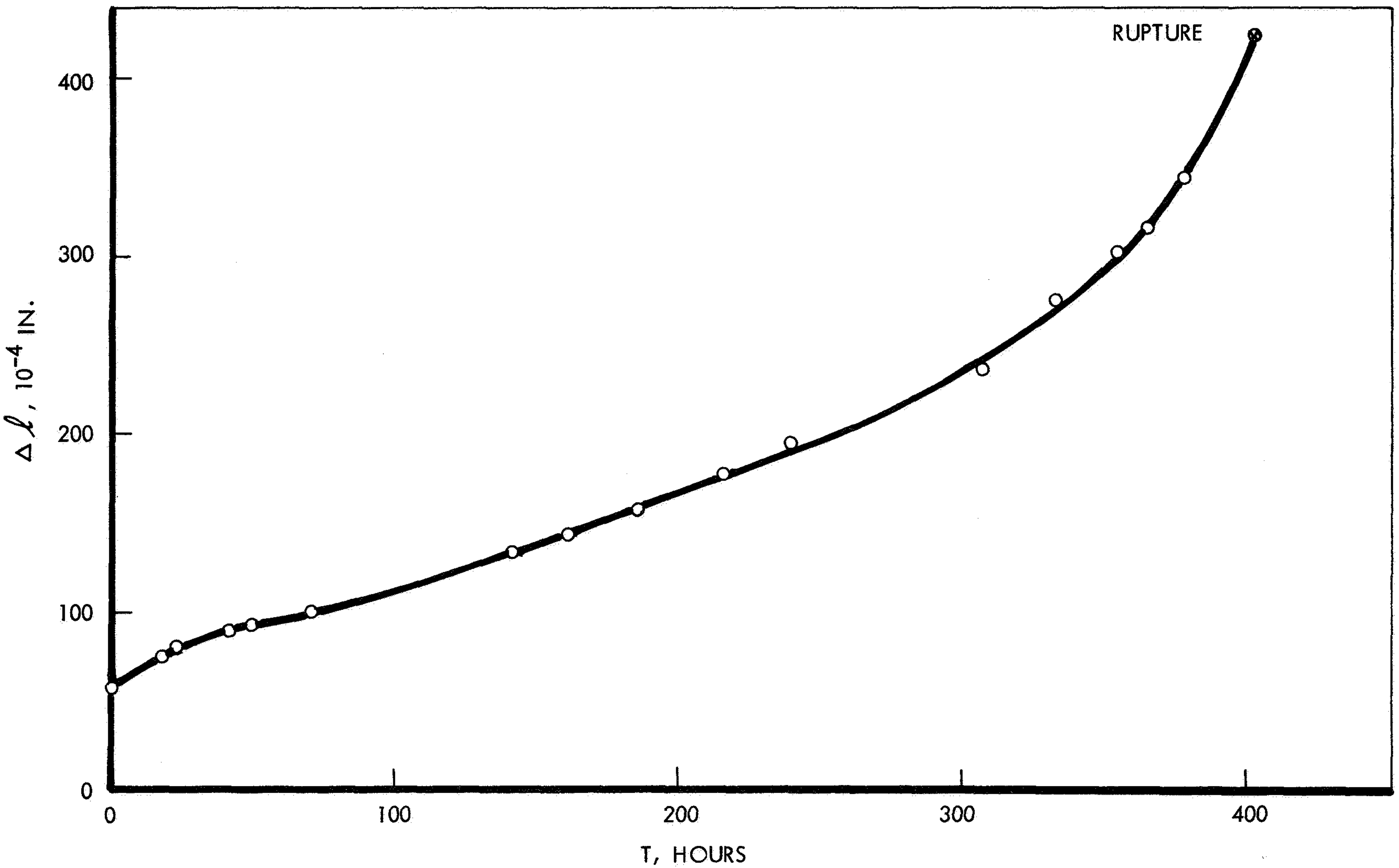


Figure 46. Creep Curve for Alloy IV Y at 1400°F Under 94,000 psi Load Using a Platinum Extensometer. Stress Rupture Life = 406.2 Hours and Percent Elongation = 3.0. Gauge Length: Radius to Radius.

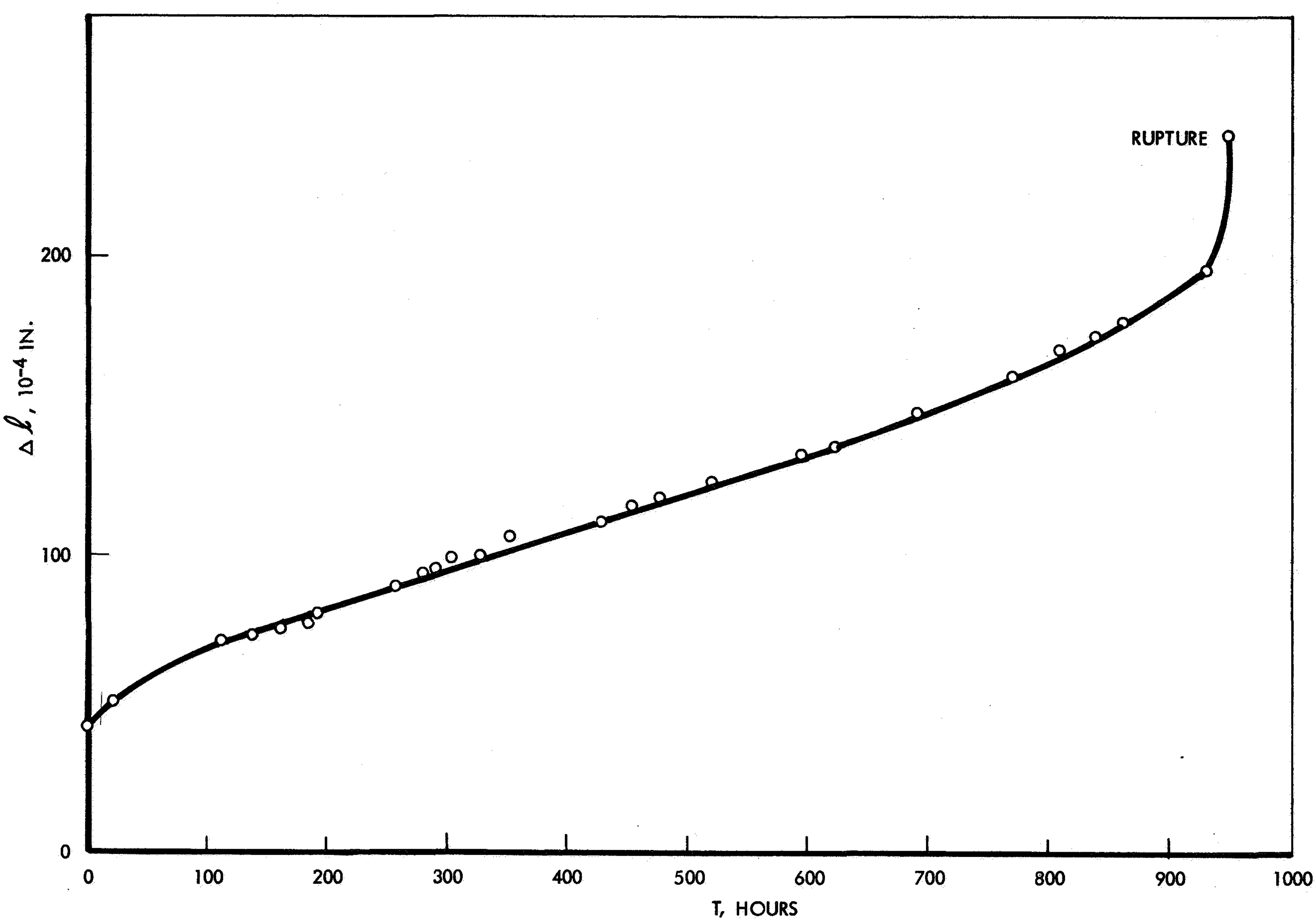


Figure 47. Creep Curve for Alloy VI A at 1400°F Under 85,000 psi Load Using a Platinum Extensometer. Stress Rupture Life = 943.6 Hours and Percent Elongation = 1.7. Gauge Length: Radius to Radius.

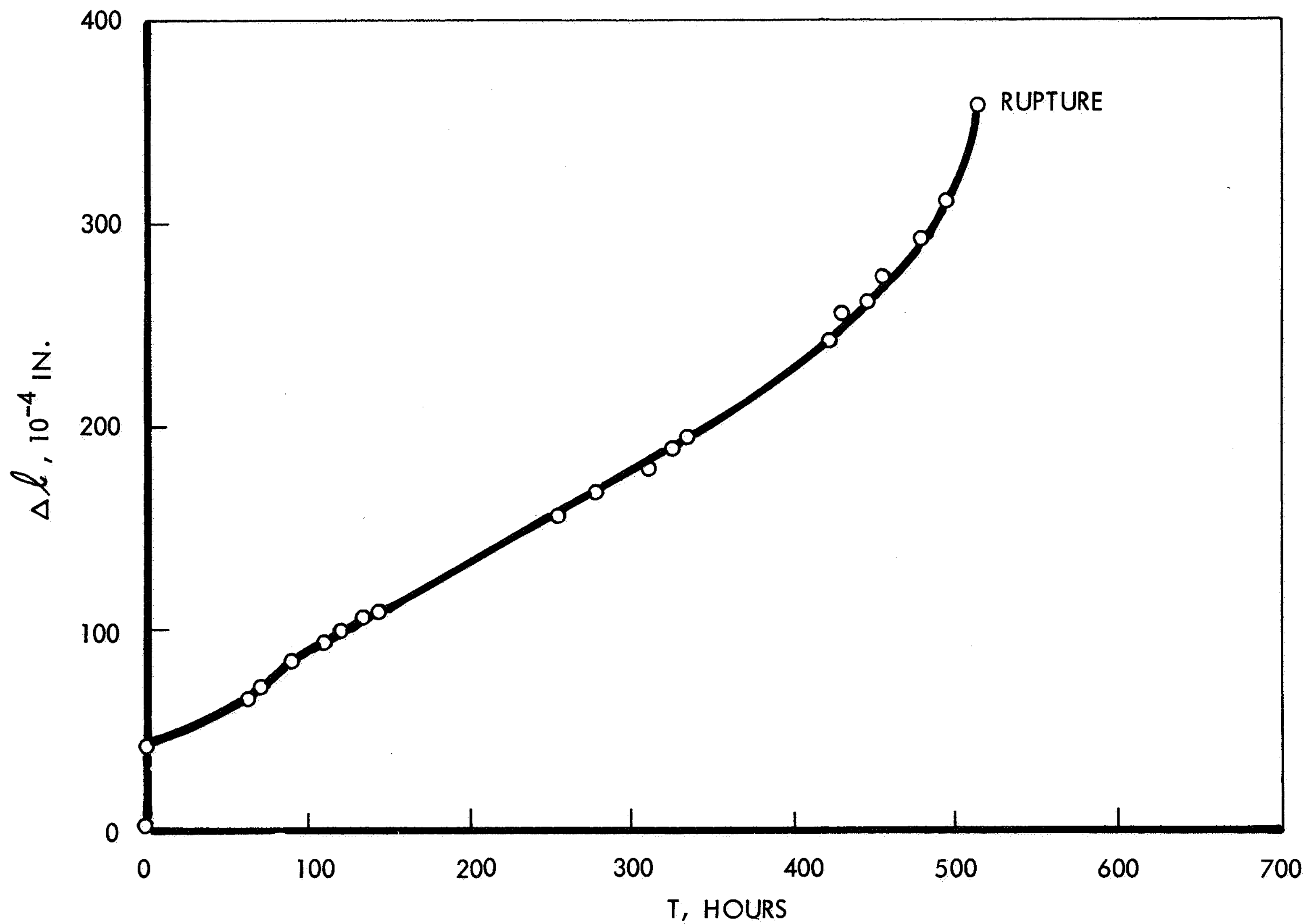


Figure 48. Creep Curve for Alloy VI A at 1400°F Under 94,000 psi Load Using a Platinum Extensometer. Stress Rupture Life = 516.7 Hours and Percent Elongation = 2.9. Gauge Length: Radius to Radius.

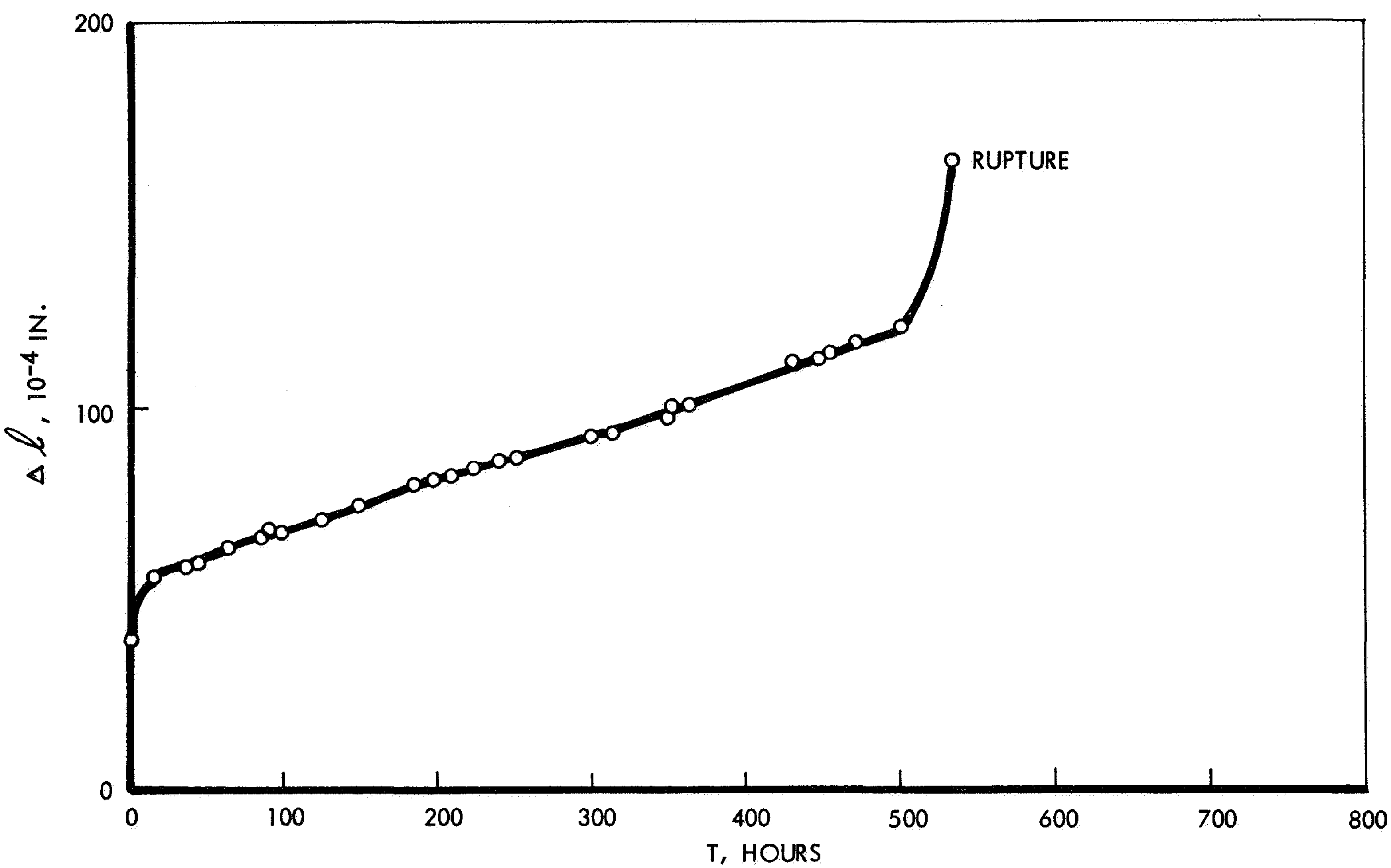


Figure 49. Creep Curve for Alloy VI D at 1400°F Under 85,000 psi Load Using a Platinum Extensometer. Stress Rupture Life = 534.9 Hours and Percent Elongation = 1.0. Gauge Length: Radius to Radius.

905 B

897 B

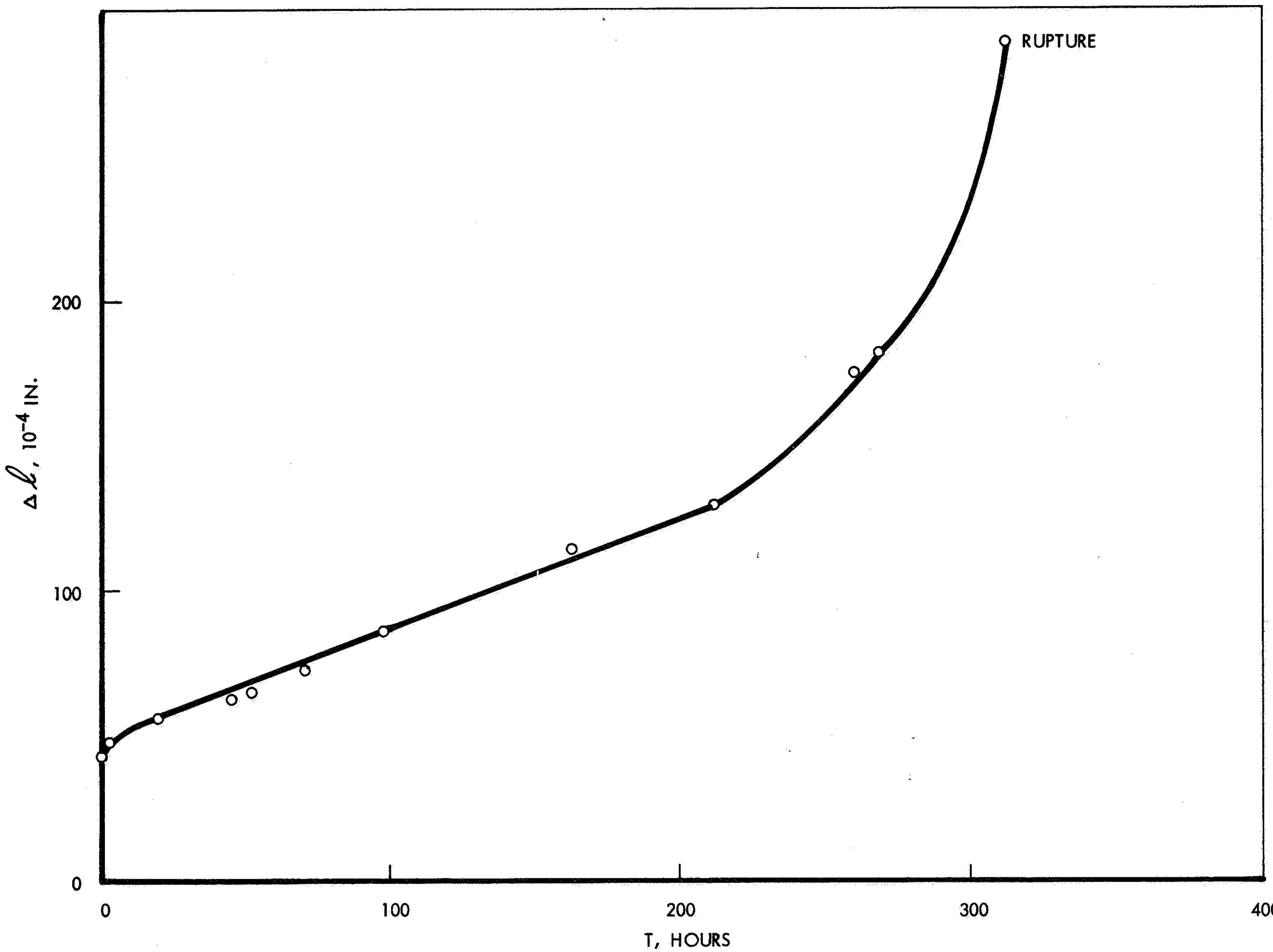


Figure 50. Creep Curve for Alloy VI D at 1400°F Under 94,000 psi Load Using a Platinum Extensometer. Stress Rupture Life = 312.4 Hours and Percent Elongation = 2.0. Gauge Length: Radius to Radius.

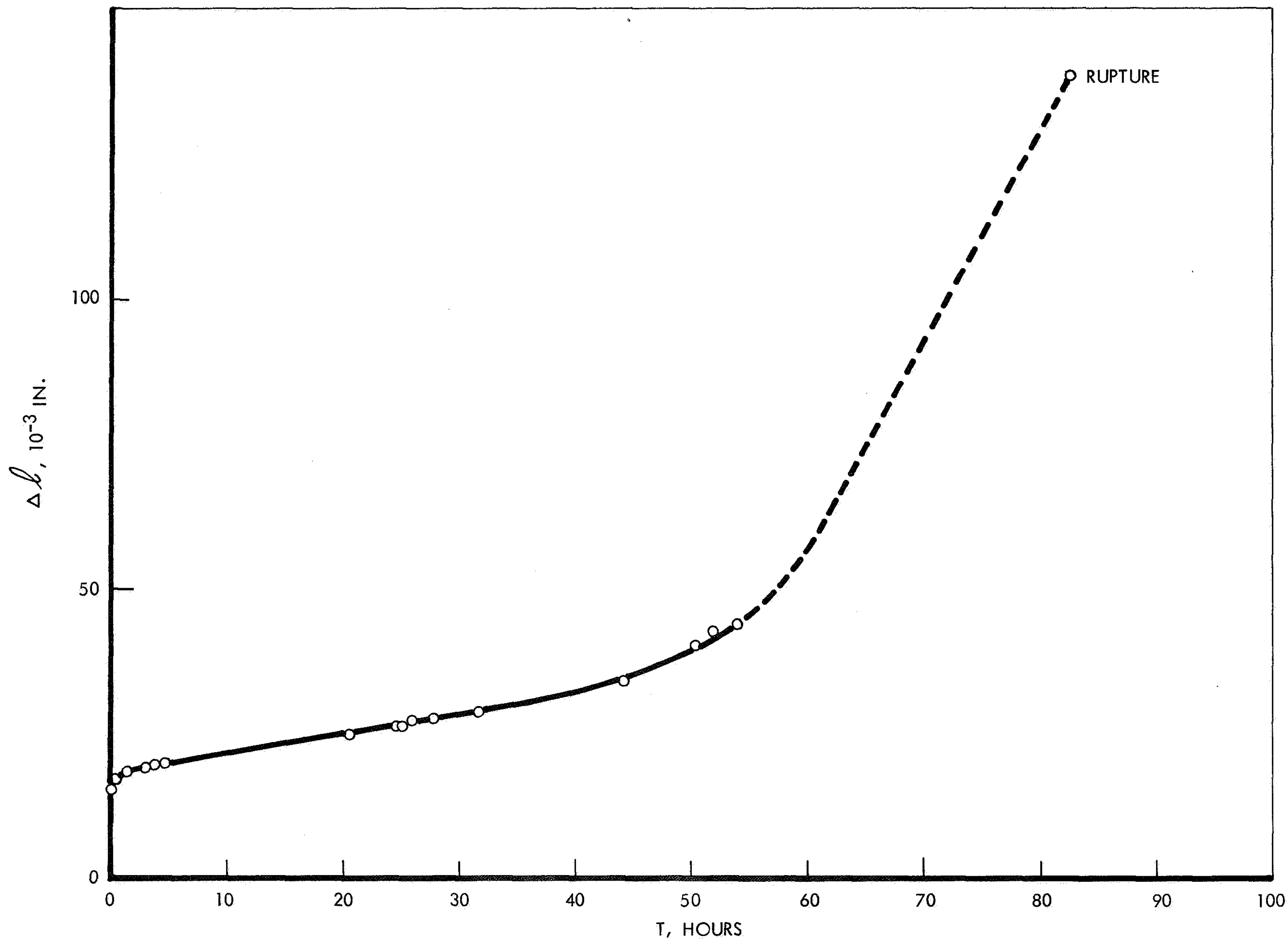


Figure 51. Creep Curve for Alloy VI A at 1875°F Under 25,000 psi Load Using a Dial Gauge. Stress Rupture Life = 82.7 Hours and Percent Elongation = 8.3. Gauge Length: Radius to Radius.

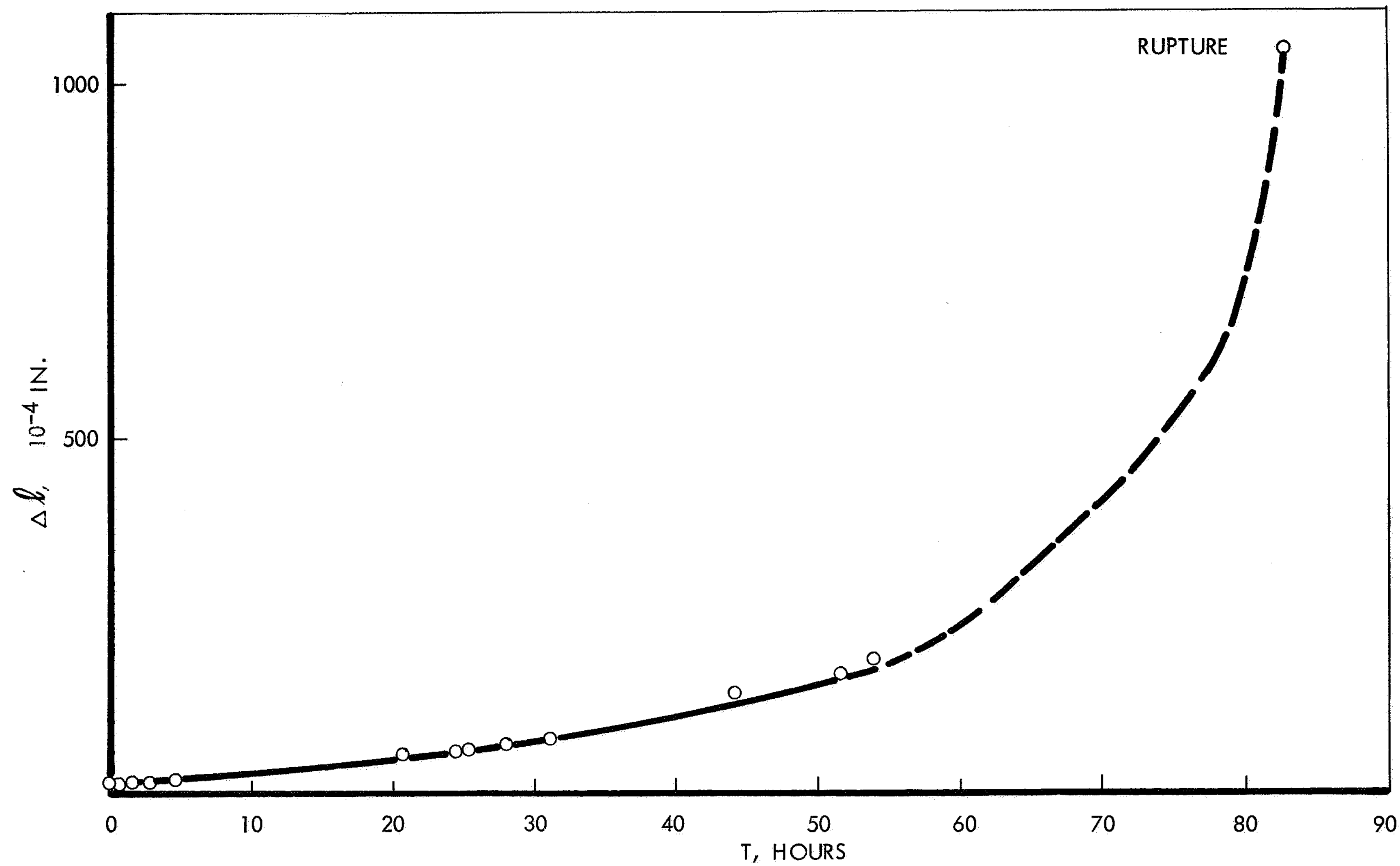


Figure 52. Creep Curve for Alloy VI A at 1875°F Under 25,000 psi Load Using a Platinum Extensometer. Stress Rupture Life = 82.7 Hours and Percent Elongation = 8.3. Gauge Length: Radius to Radius.

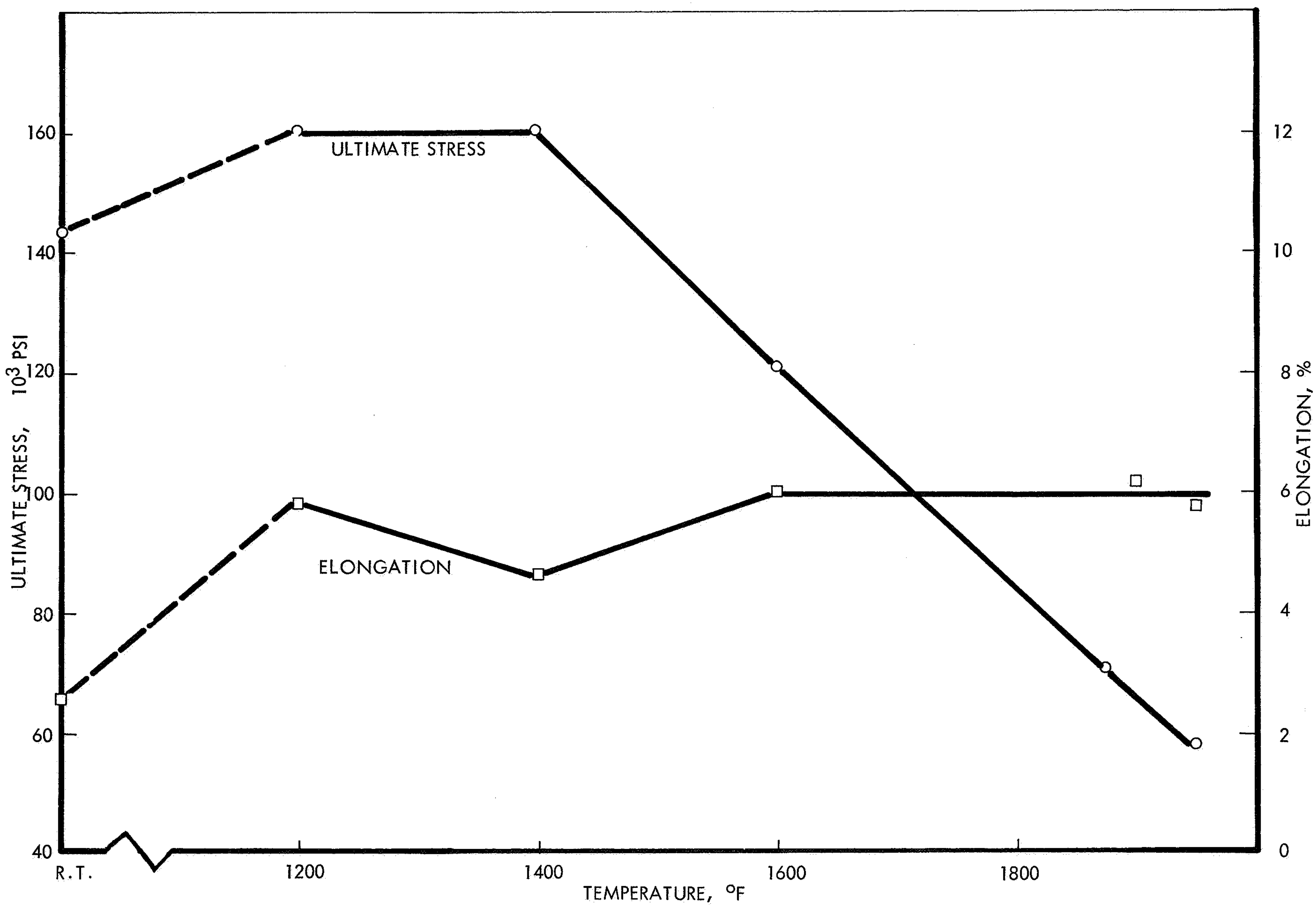


Figure 53. Tensile Properties of Alloy IV Y.

4268A

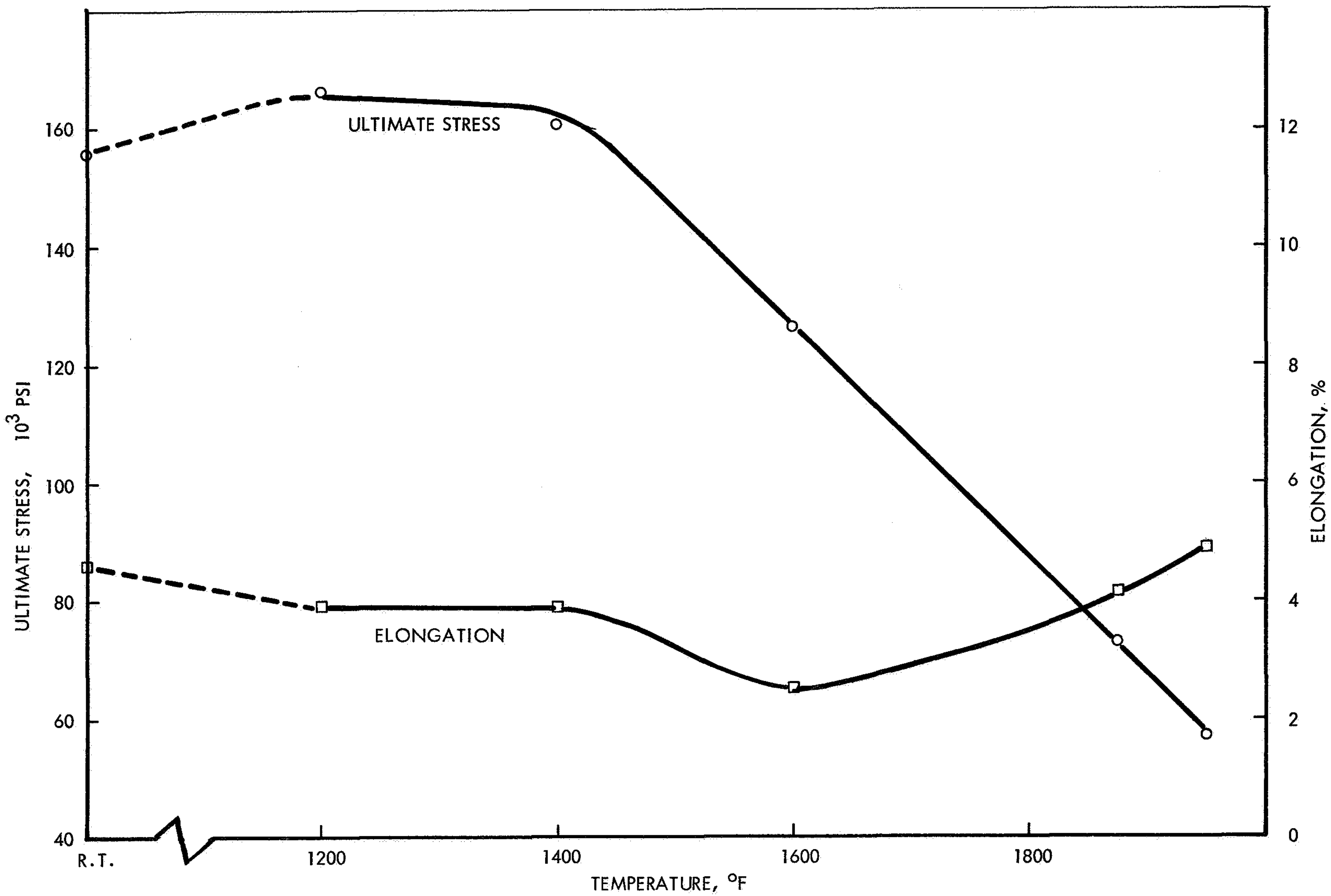


Figure 54. Tensile Properties of Alloy VI A.

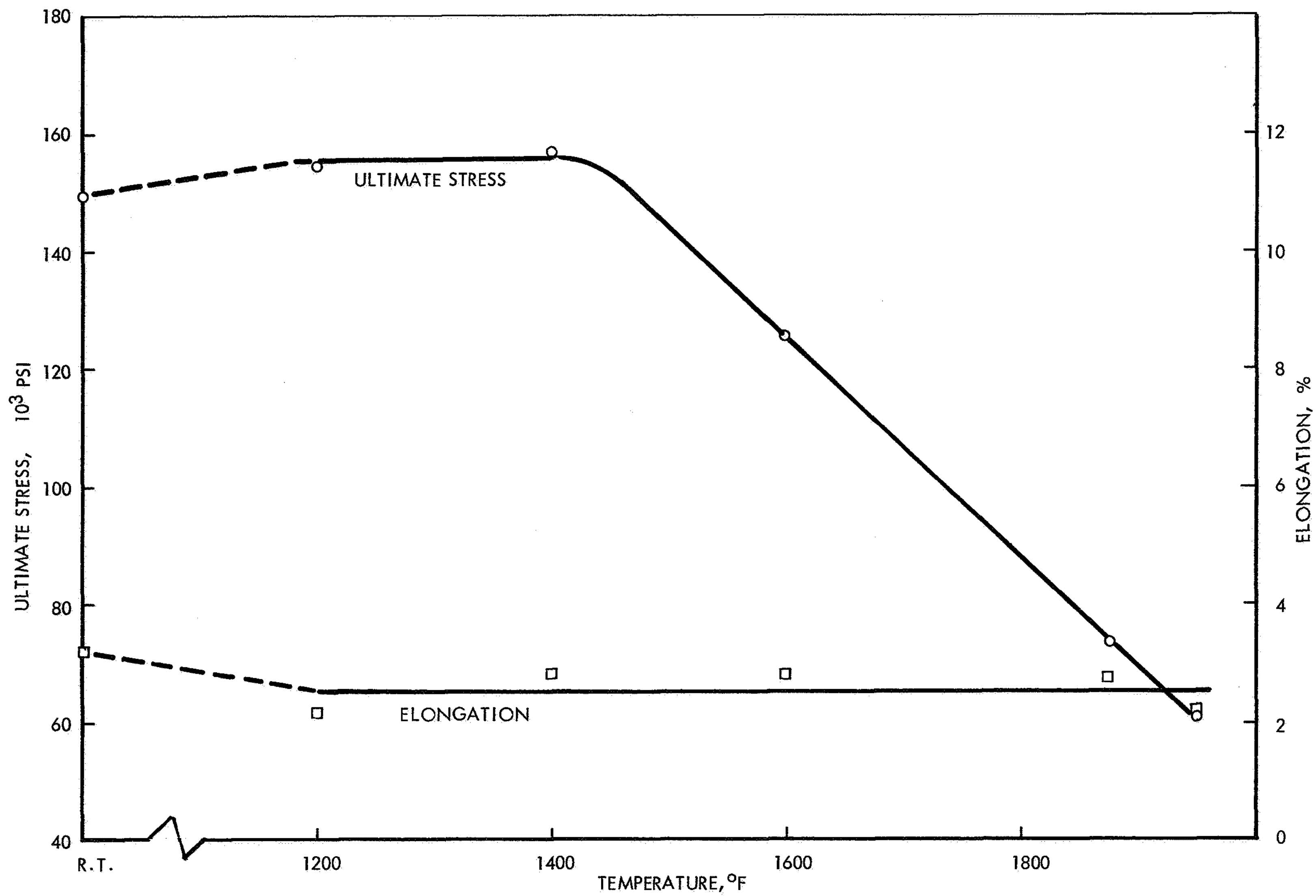
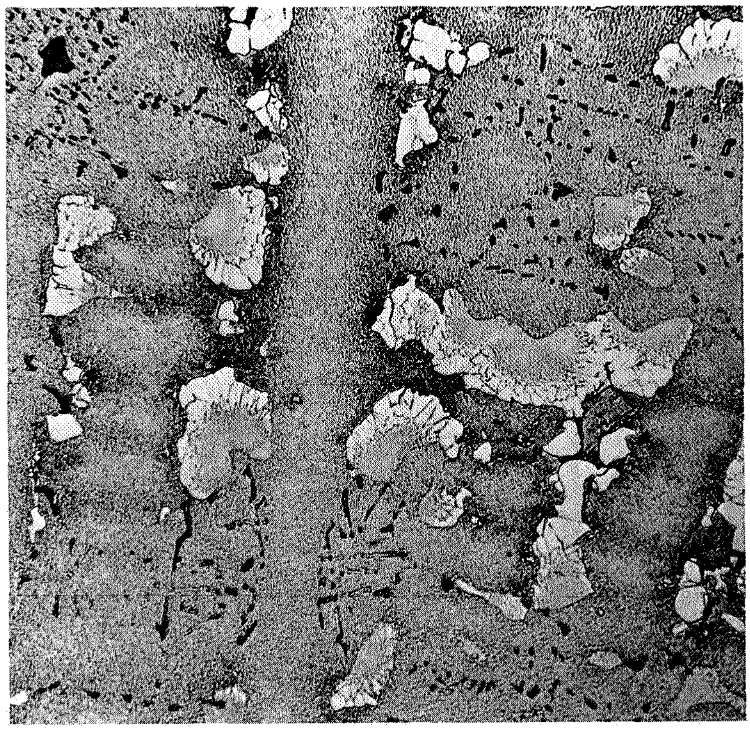
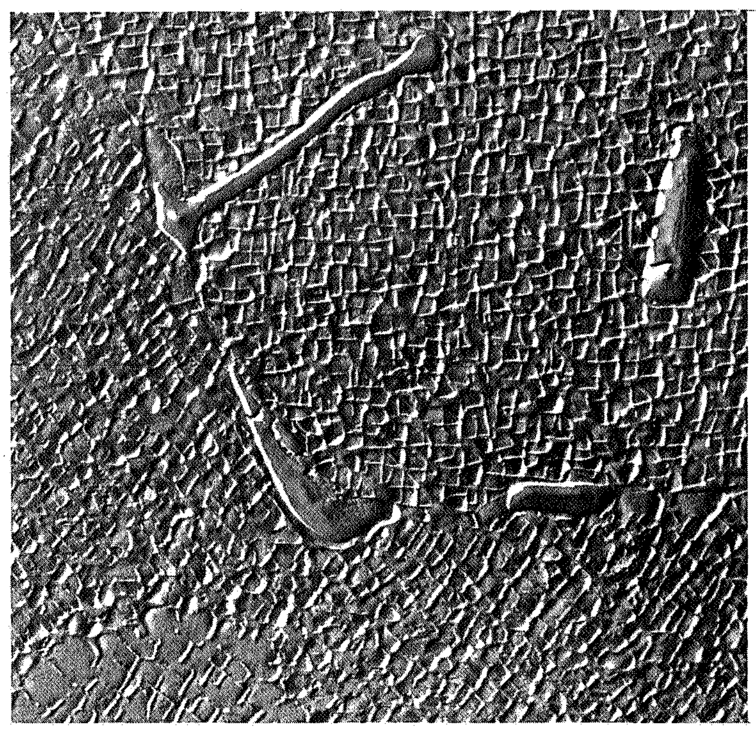


Figure 55. Tensile Properties of Alloy VI D.

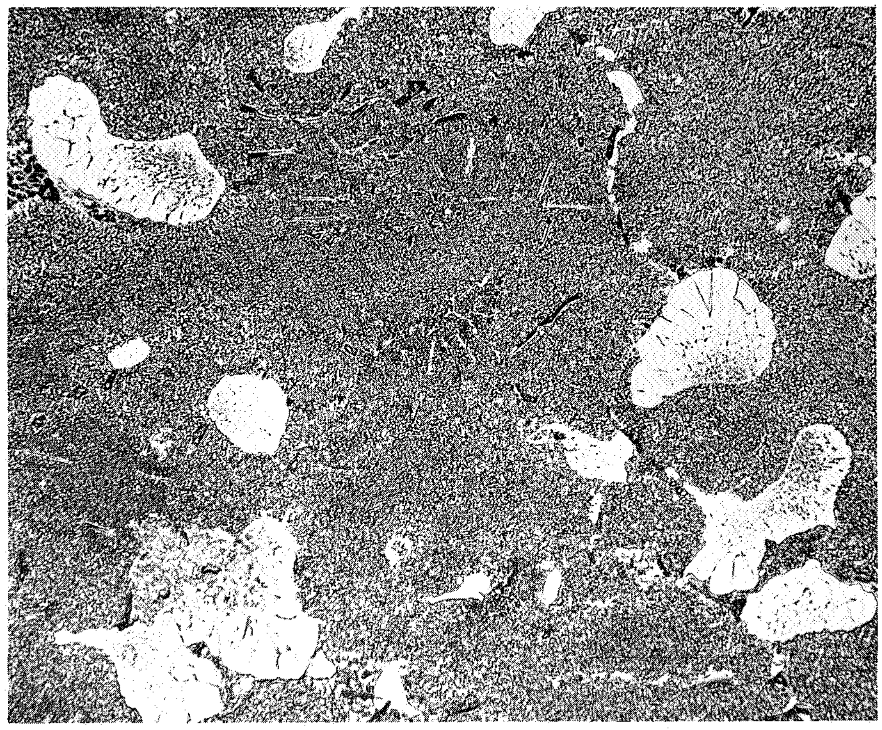


A. 250X

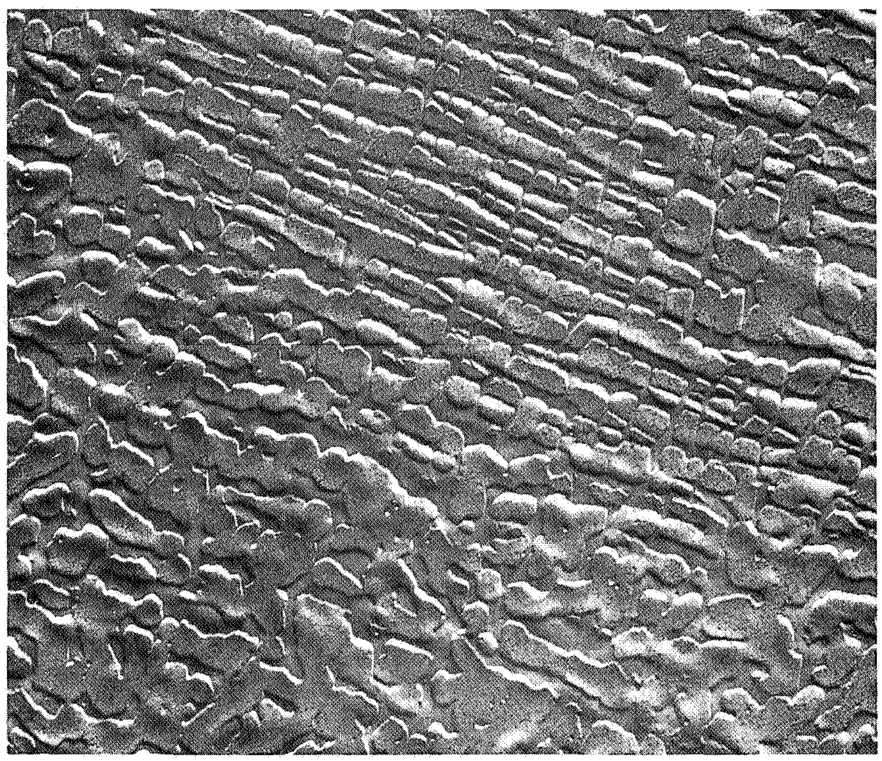


B. 4000X

Figure 56. Microstructure of Nickel-Base Alloy VI A Showing Carbide and Gamma-Prime Formations in "As-Cast" Condition. Etchant: 15ccH_2SO_4 in 100ccCH_3OH, Electrolytic Etch.

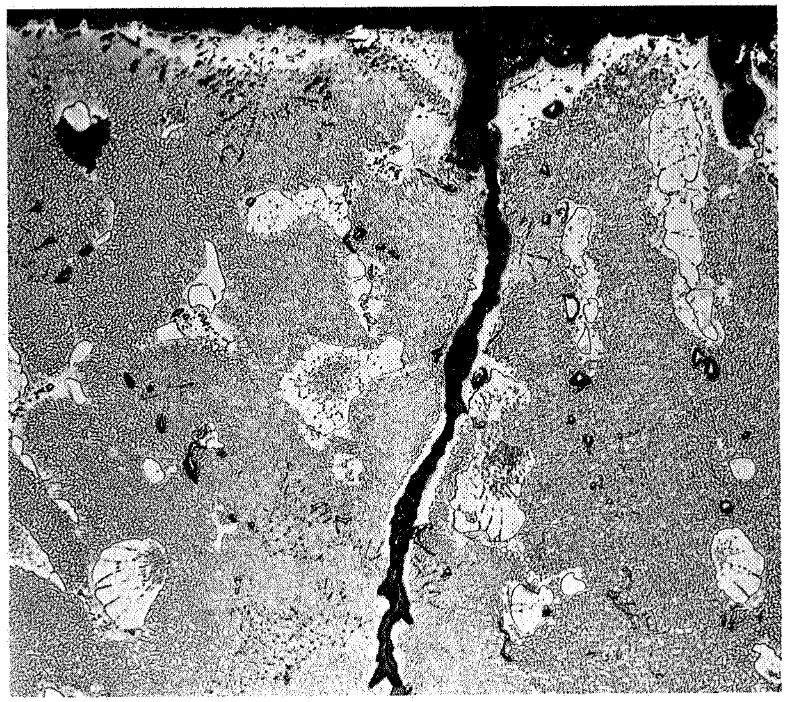


A. 250X

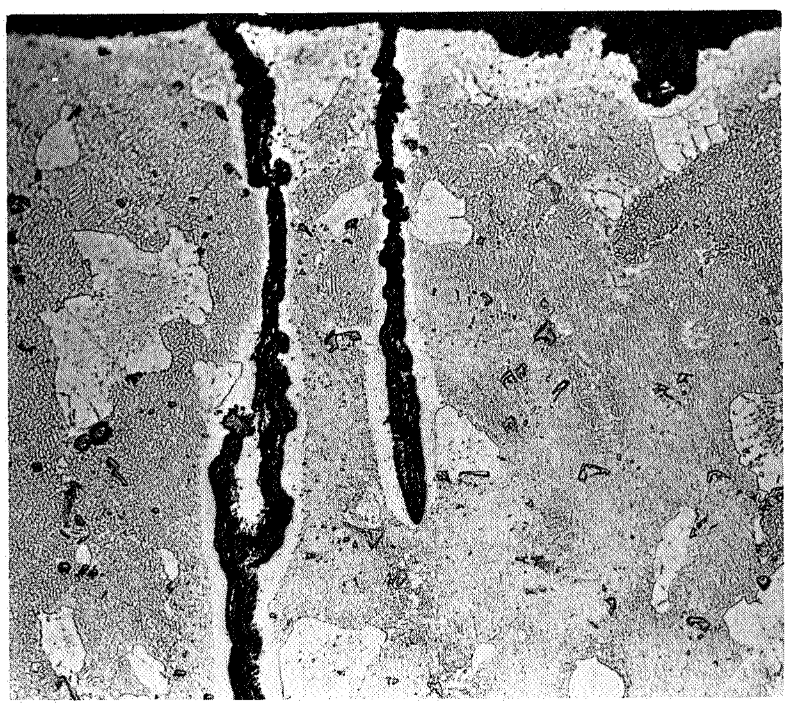


B. 4000X

Figure 57. Microstructure of Nickel-Base Alloy VI A Showing Carbide and Gamma-Prime Formations After 300 Hour Argon Age at 1875°F.
Etchant: 15cc H_2SO_4 in 100cc CH_3OH, Electrolytic Etch.

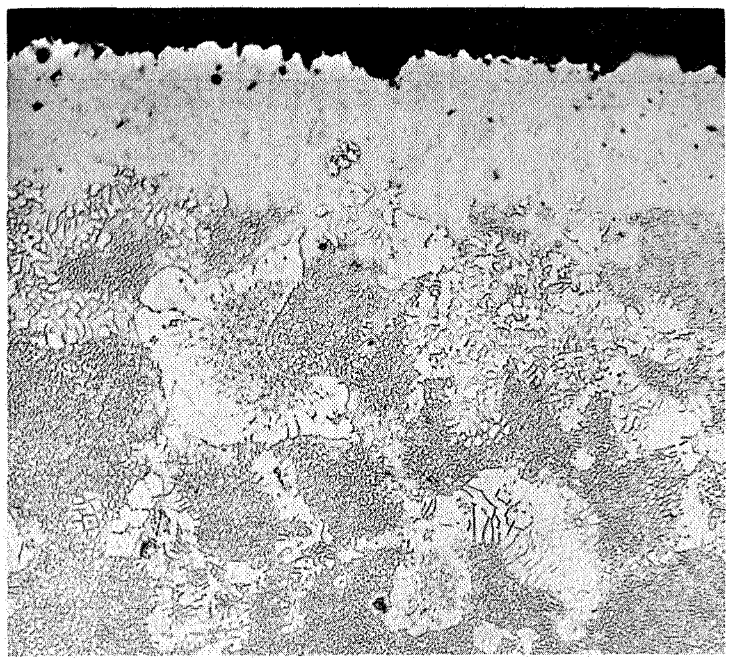


A. Alloy VI A. 250X.

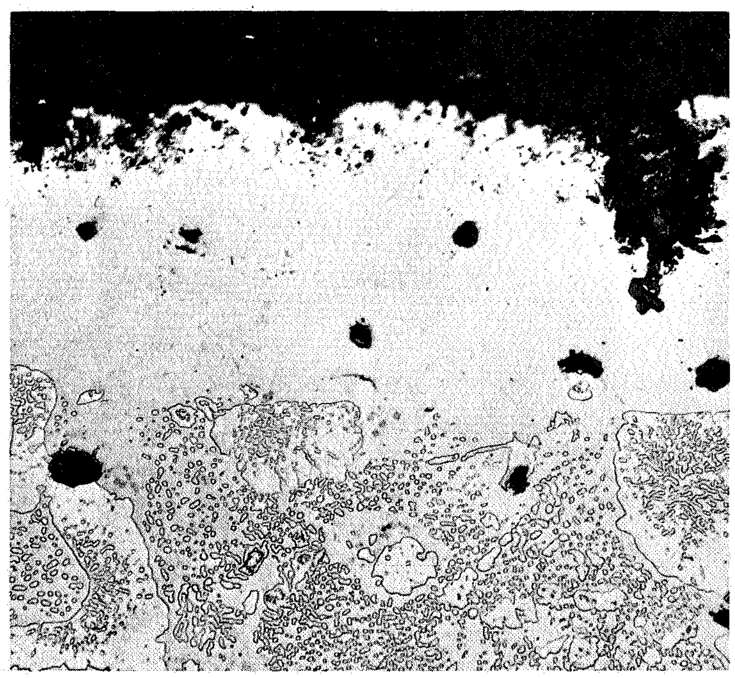


B. Alloy VI D. 250X.

Figure 58. Photomicrographs of Nickel-Base Alloys Showing Thermal Fatigue Cracks. Test Temperature: 2100°F. Etchant: 62% H_2O, 15% HF, 15% H_2SO_4, and 8% HNO_3.

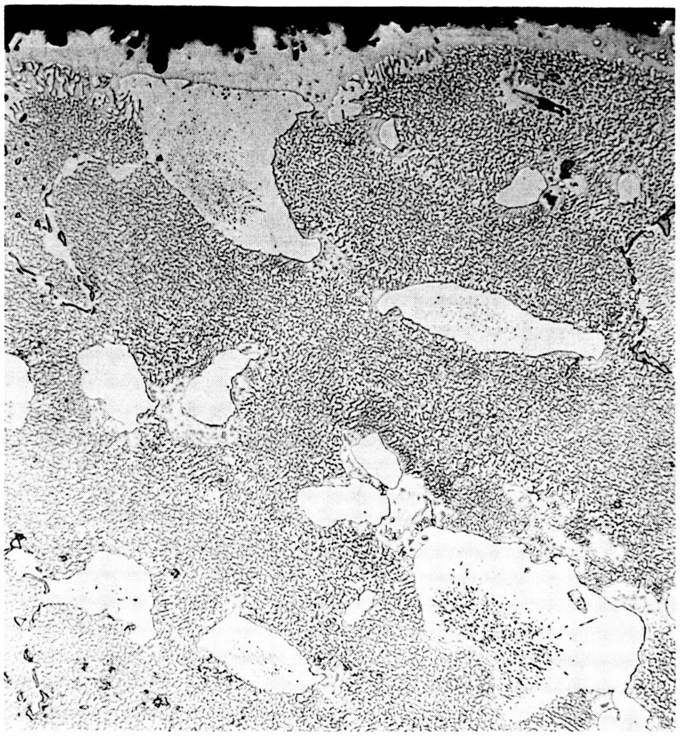


A. Test Temperature: 1875°F.

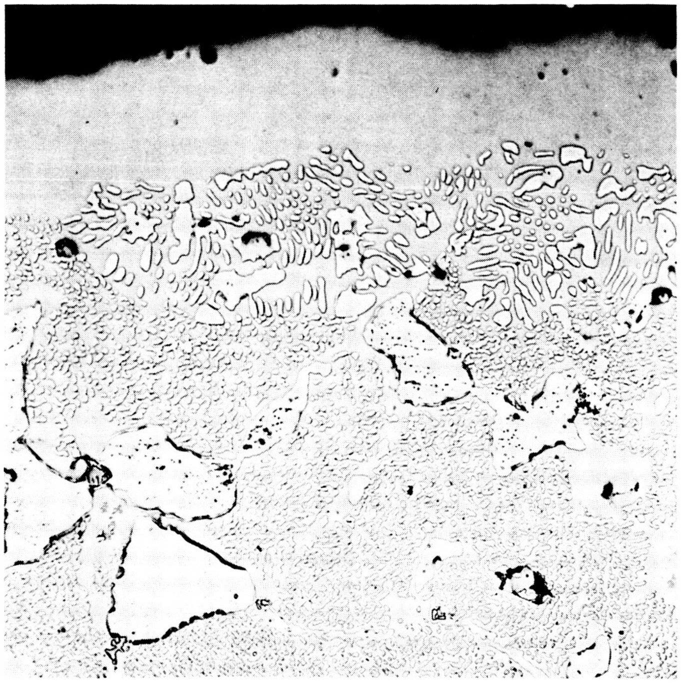


B. Test Temperature: 2100°F.

Figure 59. Photomicrographs of Oxidation Specimens of Alloy IV Y in Vicinity of Exposed Surface After an Exposure Time of 1000 Hours. 250X.
Etchant: 62% H_2O, 15% HF, 15% H_2SO_4, and 8% HNO_3.



A. Test Temperature: 1875°F.



B. Test Temperature: 2100°F.

Figure 60. Photomicrographs of Oxidation Specimens of Alloy VI A in Vicinity of Exposed Surface After an Exposure Time of 1000 Hours. 250X.
Etchant: 62% H_2O, 15 % HF, 15% H_2SO_4, and 8% HNO_3.

Figure 61. Example of Catastrophic Oxidation in Nickel-Base Alloy VI D After 500 Hours at 1950°F. 1X.

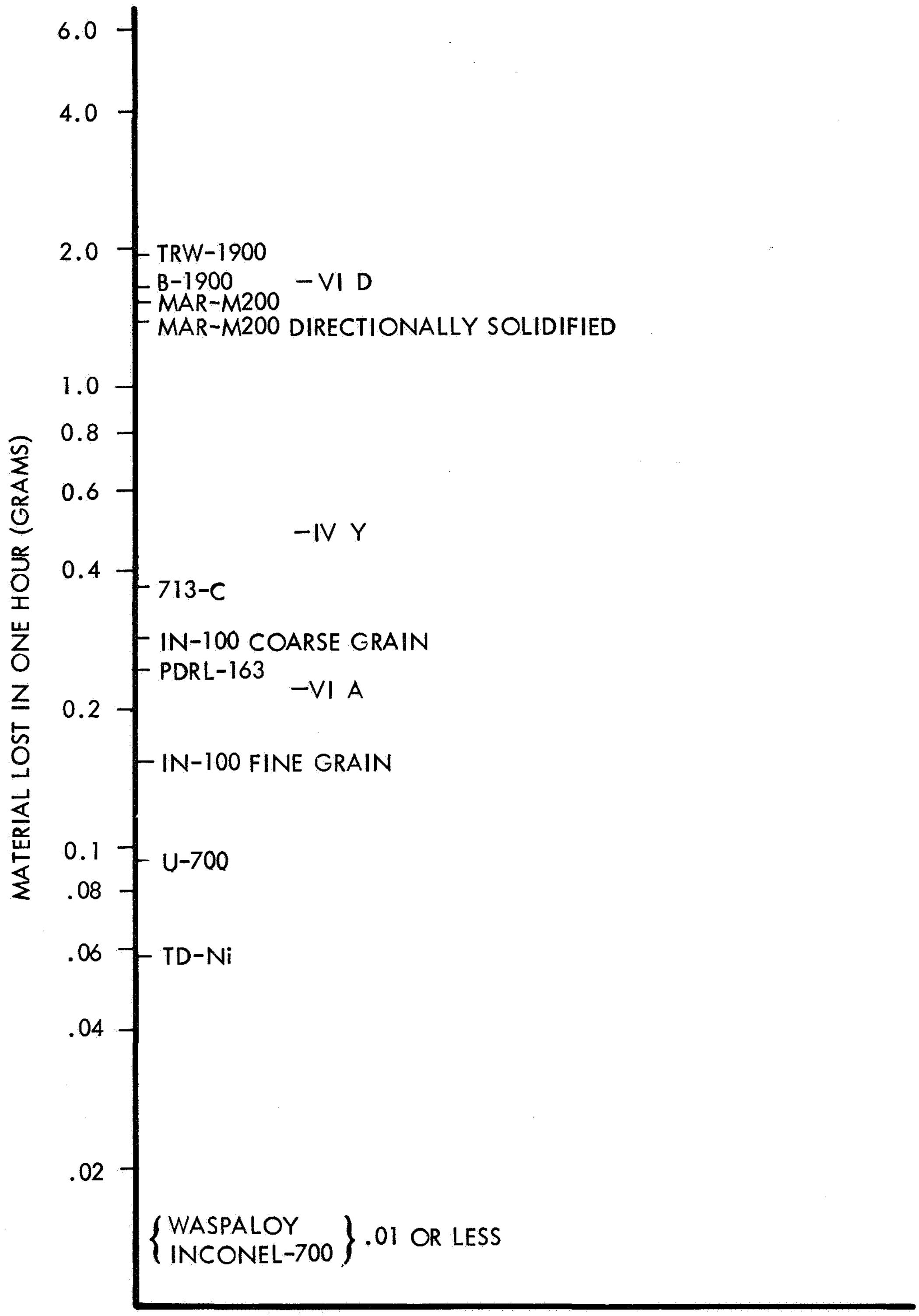


Figure 62. Comparison of Hot Corrosion Behavior at 1800°F of Nickel-Base Alloys as Determined by Crucible Test.

903 B

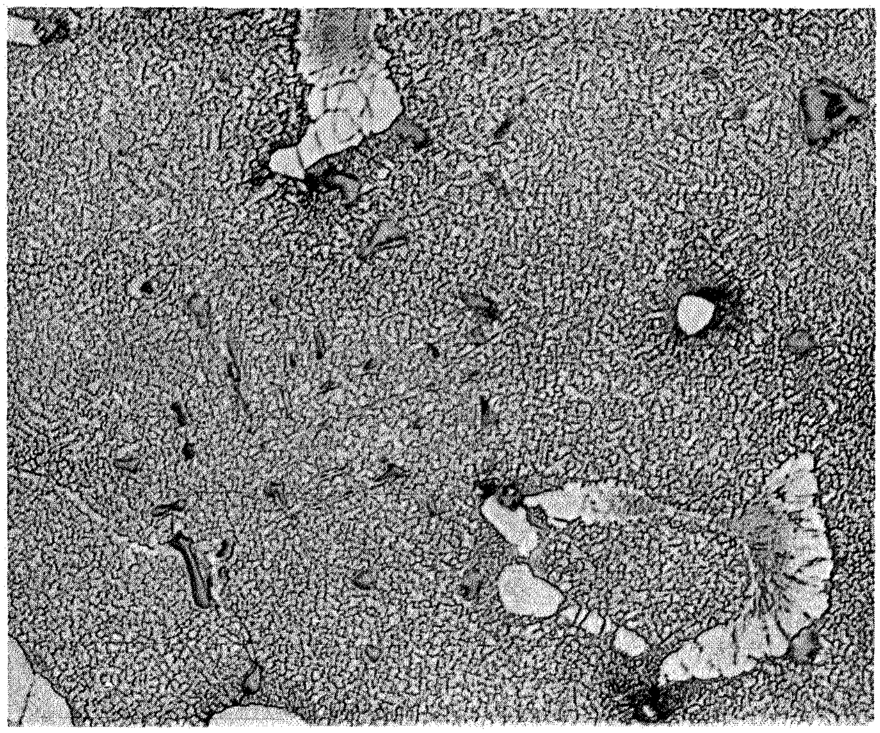
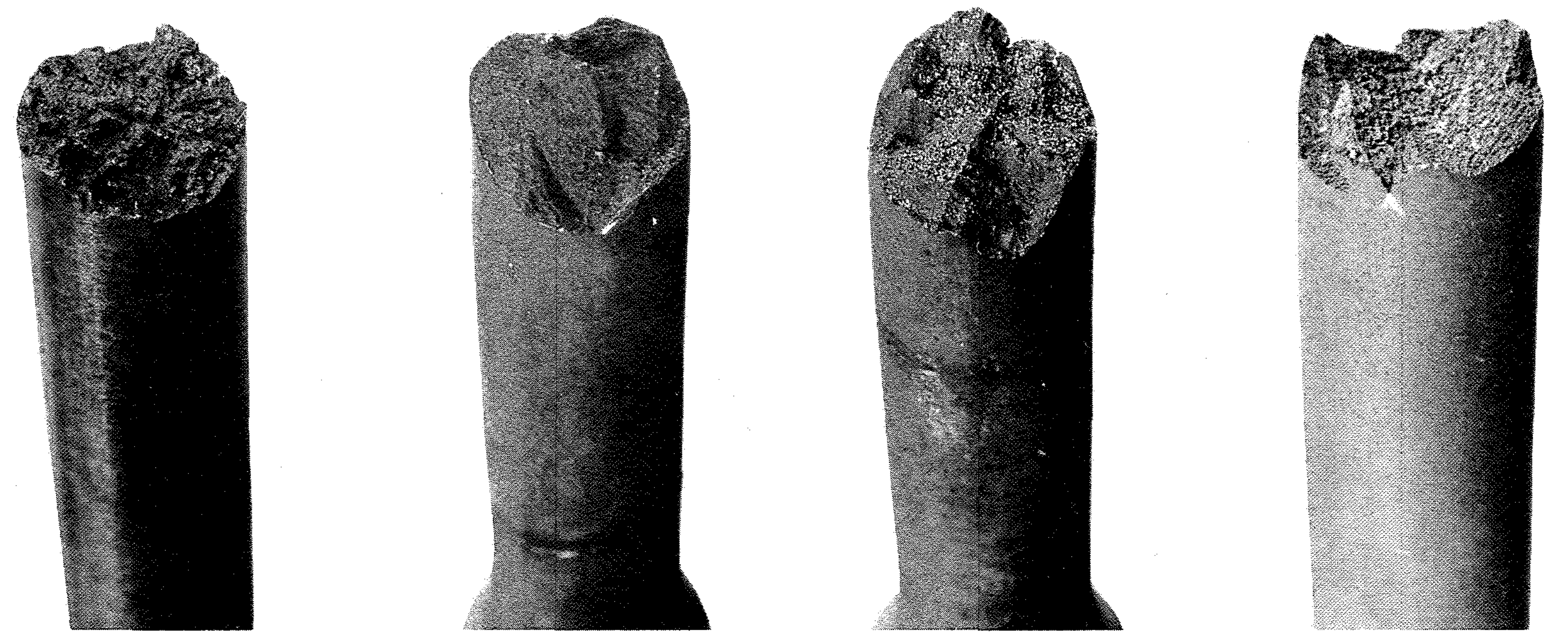


Figure 63. Microstructure of Nickel-Base Alloy VI A After a 1500 Hour Age at 1600°F. 250X.
Etchant: 62% H_2O, 15% HF, 15% H_2SO_4, and 8% HNO_3.

815 B



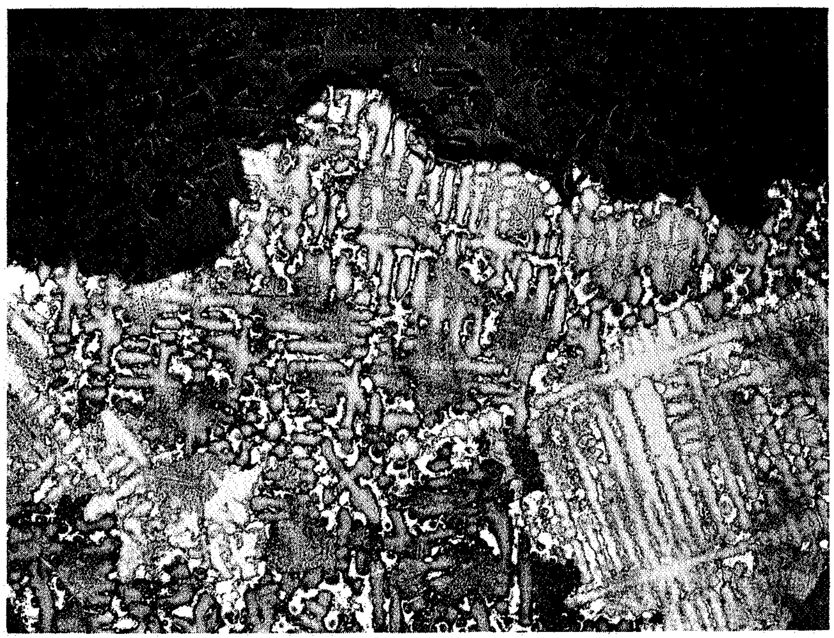
A. Alloy VI A.
$1400^{o}F/94,000$ psi
Air

B. Alloy IV Y
$1875^{o}F/15,000$ psi
Air

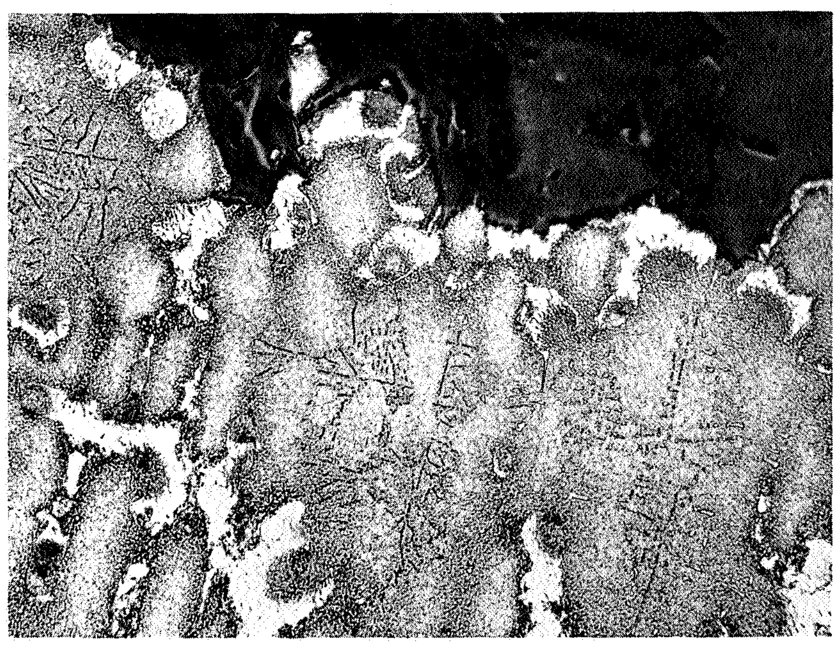
C. Alloy VI D
$1950^{o}F/15,000$ psi
Air

D. Alloy IV Y
$1950^{o}F/15,000$ psi
Argon

Figure 64. Appearance of Typical Stress Rupture Fractures in Experimental Nickel-Base Alloys. 5X.

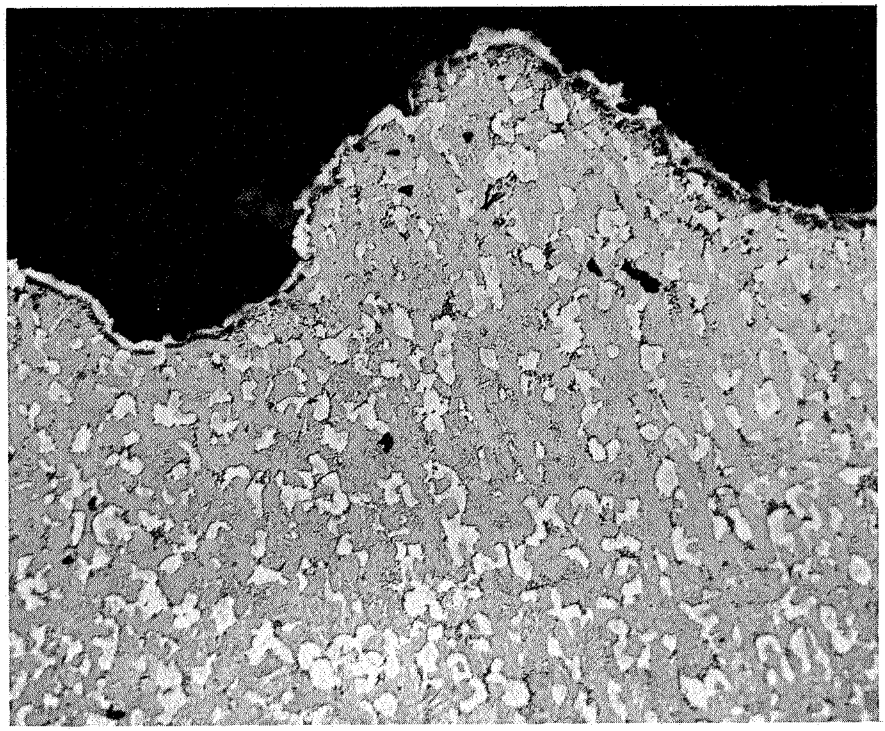


A. 60X

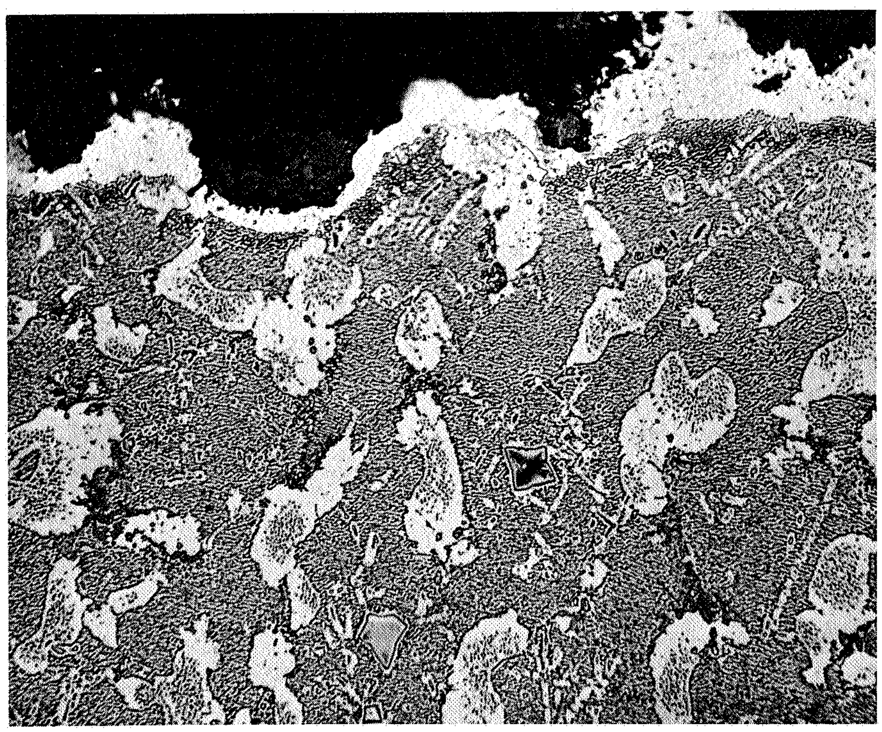


B. 250X

Figure 65. Microstructural Appearance of Nickel-Base Alloy VI A Adjacent to Fracture Showing Intergranular Nature of 1400°F/94,000 psi Stress Rupture Failure in Air.
Etchant: 62% H_2O, 15% HF, 15% H_2SO_4, and 8% HNO_3.

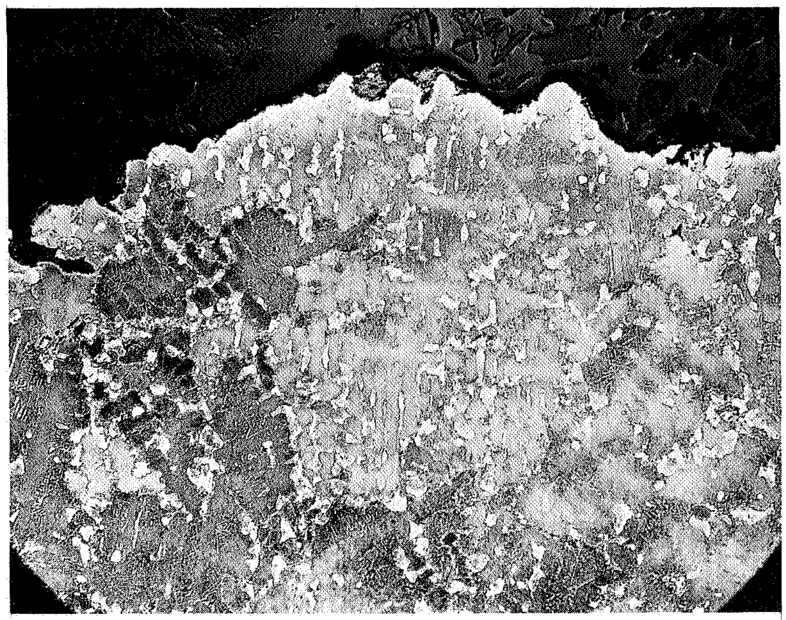


A. 60X

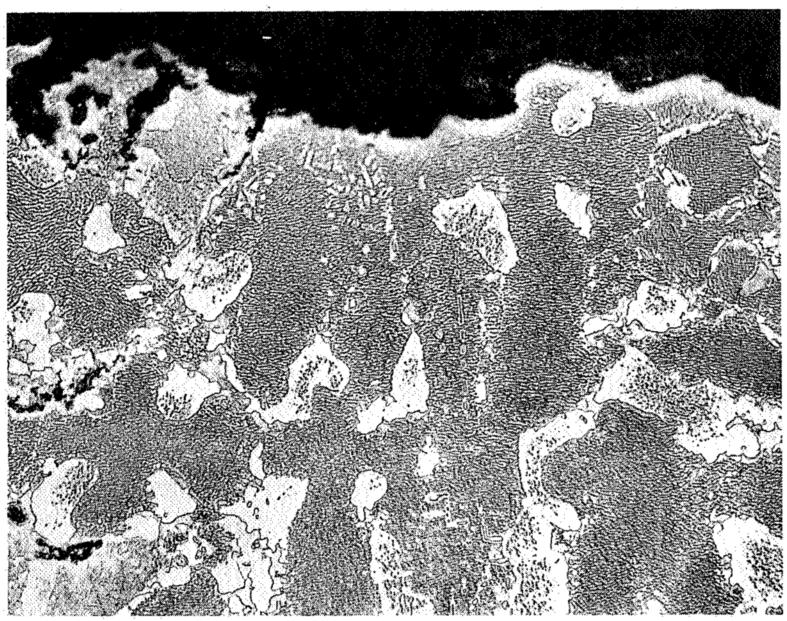


B. 250X

Figure 66. Microstructural Appearance of Nickel-Base Alloy VI D Adjacent to Fracture Showing Intergranular Nature of 1875°F/15,000 psi Stress Rupture Failure in Air. Etchant: 62% H_2O, 15% HF, 15% H_2SO_4, and 8% HNO_3.

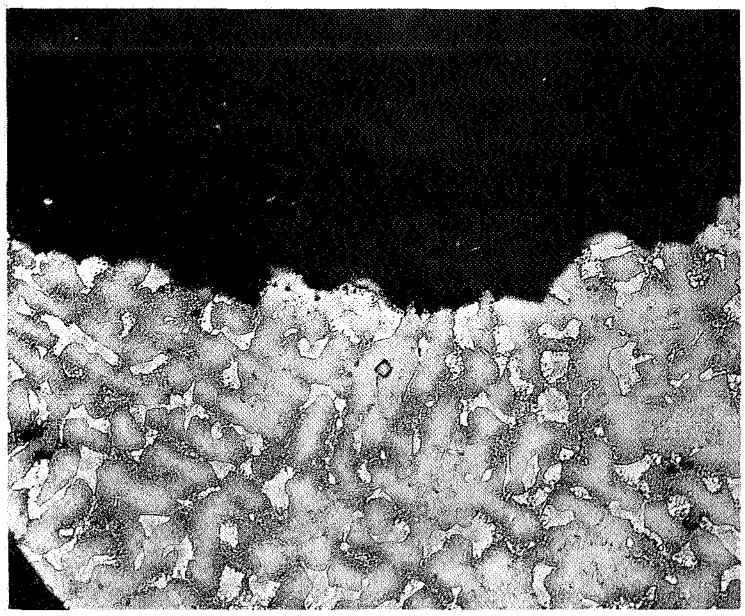


A. 60X

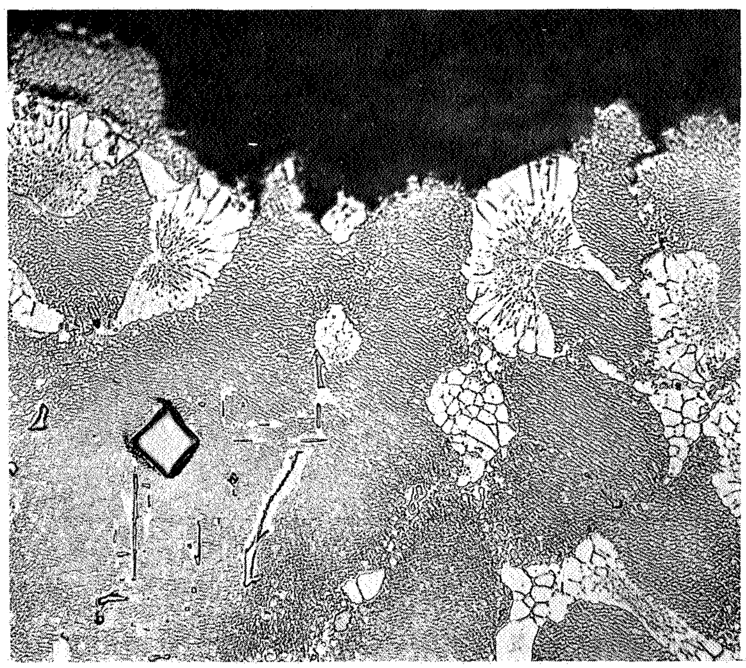


B. 250X

Figure 67. Microstructural Appearance of Nickel-Base Alloy VI A Adjacent to Fracture Showing Intergranular Nature of 1950°F/15,000 psi Stress Rupture Failure in Air.
Etchant: 62% H_2O, 15% HF, 15% H_2SO_4, and 8% HNO_3.



A. 60X



B. 250X

Figure 68. Microstructural Appearance of Nickel-Base Alloy IV Y Adjacent to Fracture Showing Intergranular Nature of $1950^{\circ}F/15{,}000$ psi Stress Rupture Failure in Argon. Etchant: 62% H_2O, 15% HF, 15% H_2SO_4, and 8% HNO_3.

824 B

A. As-Extruded Condition.

B. Heat-Treated Condition.

Figure 69. Photomicrographs of Nickel-Base Alloy IIIg Showing Typical Superalloy Material in the As-Extruded and Heat-Treated Condition. 250X.
Etchant: 62% H_2O, 15% HF, 15% H_2SO_4, and 8% HNO_3.

819 B

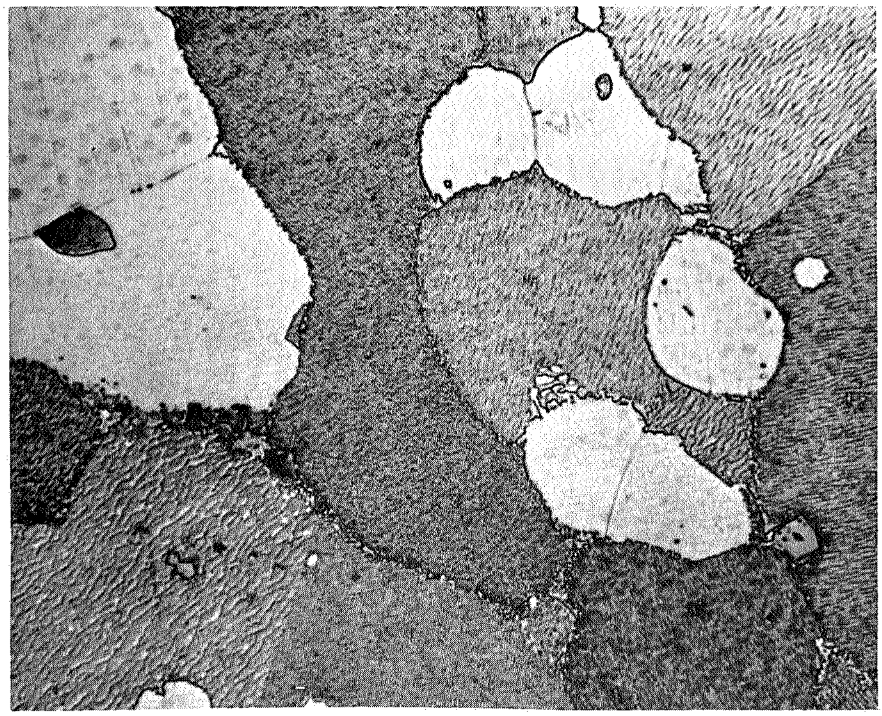
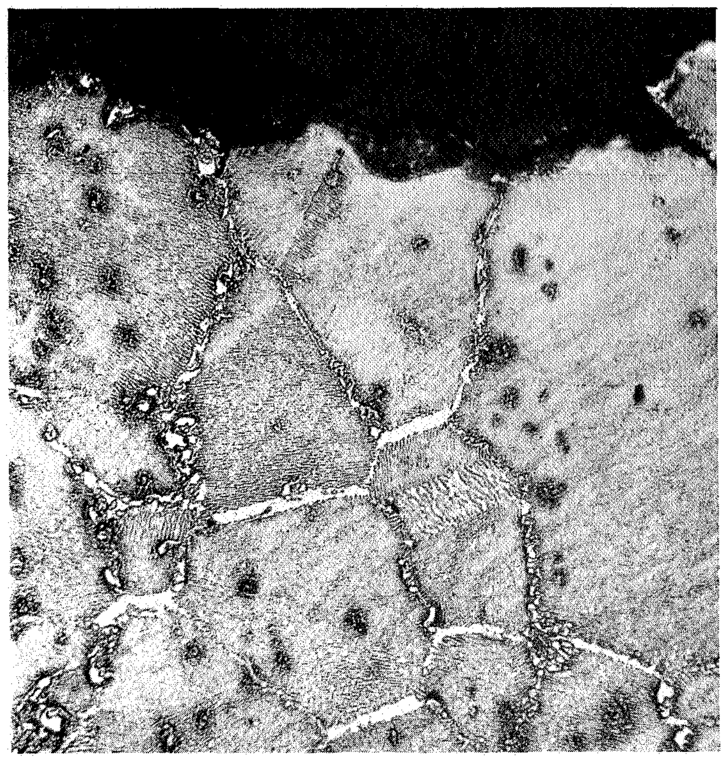


Figure 70. Photomicrograph of Nickel-Base Alloy IV Y Showing Incipient Melting in Grain Boundaries After Heat Treatment. 250X. Etchant: 62% H_2O, 15% HF, 15% H_2SO_4, and 8% HNO_3.



A. Alloy IIb, 10,000 psi Load.

B. Alloy IId, 15,000 psi Load.

Figure 71. Microstructural Appearance of Various Nickel-Base Wrought Superalloys Adjacent to Fracture Showing Intergranular Nature of 1875°F Stress Rupture Failure. 250X. Etchant: 62% H_2O, 15% HF, 15% H_2SO_4, and 8% HNO_3.

APPENDIX I

Sample Calculation for Latin Square Analysis

See Table 15 for data (log 2000°F stress rupture life x 10^2 at 15,000 psi)

Series I base with 4.5% Al and 1.0%Ti

Summation of all values $\Sigma(X)$ = 51.7518 (27 values)

Molybdenum

	Average
Sum of all 1.0% Mo results = 13.8155 (9 values)	1.5351
Sum of all 5.5% Mo results = 18.8858 (9 values)	2.0984
Sum of all 8.0% Mo results = 19.0505 (9 values)	2.1167

Sum of Squares Calculation

$$\frac{(13.8155)^2}{9} + \frac{(18.8858)^2}{9} + \frac{(19.0505)^2}{9} - \frac{(51.7518)^2}{27} = 1.9682$$

Tungsten

	Average
Sum of all 1.0% W results = 12.3781 (9 values)	1.3753
Sum of all 5.5% W results = 17.1646 (9 values)	1.9072
Sum of all 10.0% W results= 22.2091 (9 values)	2.4677

Sum of Squares Calculation

$$\frac{(12.3781)^2}{9} + \frac{(17.1646)^2}{9} + \frac{(22.2091)^2}{9} - \frac{(51.7518)^2}{27} = 5.3706$$

Tantalum

	Average
Sum of all 1.0% Ta results = 10.2031 (9 values)	1.1337
Sum of all 4.5% Ta results = 14.8696 (9 values)	1.6522
Sum of all 8.0% Ta results = 26.6791 (9 values)	2.9643

Sum of Squares Calculation

$$\frac{(10.2031)^2}{9} + \frac{(14.8696)^2}{9} + \frac{(26.6791)^2}{9} - \frac{(51.7518)^2}{27} = 16.0258$$

Total

Sum of squares of all results $\Sigma(X^2)$ = 123.2497

Subtracting from this value the sum of all values $\frac{(\Sigma X)^2}{27}$ = 99.1944

Total = $\Sigma(X^2) - (\Sigma X)^2/27$ = 24.0553

Degrees of Freedom

Degrees of freedom for each Mo, W and Ta = N-1 = 2
(where N is the number of tests conducted)

Degrees of freedom of total = N-1 = 27-1=26

Residual and Interaction

Degrees of freedom = 26-6=20

Sum of squares of residual and interaction terms =

Total - (sum of other sums of squares)

24.0553 - (1.9682 + 5.3706 + 16.0258 = 0.6907

Mean Square

Mean square of Mo = sum of squared $\div$ degrees of freedom
= 1.9682 $\div$ 2 = 0.9841

Tabular Form

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio*
Mo	1.9682	2	0.9841	28.52
W	5.3706	2	2.6853	77.83
Ta	16.0258	2	8.0129	232.26
Residual and Interaction	0.6907	20	0.0345	
Total	24.0553	26		

* The F Ratio = $\frac{\text{Mean Square of the Variable}}{\text{Mean Square of Residual and Interaction}}$

Consulting a table of F ratios,

F ratio (2,20) for 99.0% significance $\geq$ 5.85

Therefore, for this example all variables are considered significant.

APPENDIX II

Sample Calculation for Fractional Factorial Experiments

The percent elongation data of 2000°F stress rupture tests from Series IV (Table 33) will be used to illustrate the Yates' method[16] of obtaining estimates of main effects and interactions for a two-level fractional factorial experiment. The fractional factorial experiment of Series IV was designed for 16 observations with 8 elements varied at two levels (n=8) and 1/16 replication ($1/16 = 1/2^4 = 1/2^b, \therefore b = 4$), i.e., only 1/16 of the number of observations required for a full factorial experiment (256 observations) are made. The procedure for the Yates' method is as follows:

(1) A table of n' + 2 columns is made where n' = n - b. In this example, n' = 4 and so n' + 2 = 6 columns, Table B1. The treatment combinations are listed in standard order in column 1, i.e., the order listed in Table 4. (For other fractional factorial plans, see Table 12-4 of Natrella[16] or Davies[17] for the necessary ordering for the Yates' method of analysis.)

(2) The observed response or, in this example, the average % elongation of 2000°F stress rupture tests corresponding to each treatment combination listed in column 1 is entered in column 2.

(3) In the top half of column 3, the sums of consecutive pairs of entries in column 2 are entered in order, e.g., 3.97 + 2.27 = 6.24, 5.73 + 7.27 = 13.00, etc. In the bottom half of the column, the differences between the same consecutive pairs of entries, i.e., second entry minus first entry, fourth minus third, etc., are entered in order, e.g., 2.27 - 3.97 = -1.70, 7.27 - 5.73 = 1.54, etc.

(4) The remaining columns are obtained in the same manner as column 3, i.e., by obtaining in each case the sums and differences of the pairs in the preceding column in the manner described in step 3.

(5) The entries in the last column (column n' + 2) correspond to the ordered or estimated effects. Estimates of the main effects and the interactions are obtained by dividing the appropriate entry in the last column by $2^{n'-1}$. The entries in the last numbered column of Table A1 (column 6) are divided by $2^{4-1} = 2^3$ or 8 to obtain these estimates, e.g., 69.52/8 = 8.699, 4.44/8 = 0.554, etc. Two additional unnumbered columns were added to Table A1 to show the effects and the interactions being estimated and the results for this example.

APPENDIX III

Selection of Nickel-Base Alloy VI A

The quadratic equation from which the composition of Alloy VI A was formulated was obtained in the following manner: Series V alloys were designed so that they would augment the Series IV alloys in such a way that a regression equation of the form

$$\begin{array}{llllll} y = a_0 & + a_1x_1 & + a_2x_2 & + a_3x_3 & + a_4x_4 & \\ & + a_{11}x_1^2 & + a_{12}x_1x_2 & + a_{13}x_1x_3 & + a_{14}x_1x_4 & \\ & & + a_{22}x_2^2 & + a_{23}x_2x_3 & + a_{24}x_2x_4 & \\ & & & + a_{33}x_3 & + a_{34}x_3x_4 & \\ & & & & + a_{44}x_4 & \quad \text{(A1)} \end{array}$$

could be fitted, where y is a measured property (effect or response), the a's are regression coefficients, and each x is the weight percent of an alloying element designated by its index. Only four elements were varied in Series V; namely, Cr, Ta, W, and Hf. All of the rest were fixed at their levels corresponding to the Series IV design-center (Alloy IV Z). These four elements were included in Equation A1 as independent variables. The property studied was 2000°F stress rupture life (in logarithmic form). (Note, however, that any one of the other properties studied could be treated in a similar manner.)

Sixteen of the alloys from Series IV (i.e., those forming the fractional factorial design) and all eight from Series V were included in the regression analysis. The independent variables (x's) of Equation A1 are defined in the following manner: x_1 = (Cr), x_2 = (Ta), x_3 = (W), and x_4 = (Hf). The data for each observation includes, along with the value of each independent variable, the corresponding value of each dependent variable. A regression run treats an arbitrary number of dependent variables, calculating the coefficients of the final regression equation* for each dependent variable in sequence. The analysis resulted in the following equation:

$$\begin{array}{llll} \multicolumn{4}{l}{\text{Log}_{10}\ (2000°\text{F stress rupture life, hours}) = -7.088} \\ +0.604(\text{Cr}) & + 1.06(\text{Ta}) & + 0.70(\text{W}) & + 0.34(\text{Hf}) \\ -0.0301(\text{Cr})^2 & - 0.0226(\text{Cr})(\text{Ta}) & -0.0054(\text{Cr})(\text{W}) & -0.006(\text{Cr})(\text{Hf}) \\ & - 0.044(\text{Ta})^2 & -0.021(\text{Ta})(\text{W}) & -0.026(\text{Ta})(\text{Hf}) \\ & & -0.041(\text{W})^2 & -0.006(\text{W})(\text{Hf}) \\ & & & -0.041(\text{Hf})^2 \quad \text{(A2)} \end{array}$$

(Equation A2 is Equation 1 in the Discussion Section.)

* Note that the computer program utilized in this work uses a "stepwise" technique which generates many incomplete regression equations as intermediate steps in the calculation of a final regression equation having a full set of non-vanishing regression coefficients. Therefore, for exactness, it is necessary to refer always to the <u>final</u> regression equation.

The above regression equation was investigated further to find the location and characteristics of the stationary point. The equation was differentiated with respect to each weight percent. The resulting (linear) expressions were set to zero and solved simultaneously for the weight percents. These values of the weight percents define the location of the stationary point. Substituting them back into the regression equation gives the predicted response at that point. The stationary point for Equation A2 was located at: (Cr)=6.10, (Ta)=8.96, (W)=5.81, and (Hf)=0.432.

Note that all the quadratic coefficients are present with the same algebraic sign in the above solution. Since that sign is negative, a true maximum in the response surface is indicated without carrying out a canonical transformation, i.e., a transformation to a coordinate system in which the coefficients of all cross products, such as (Cr)(W), vanish.

The selection of Alloy VI A corresponded to the aforementioned stationary point (maximum) derived from Equation A2. Cr, Ta, W, and Hf were fixed at 6.1, 9.0, 5.8, and 0.43 respectively. All of the other alloying elements were fixed at their Series IV design-center levels (Alloy IV Z). The predicted value of 2000°F stress rupture life can be obtained by substituting the above weight percents in Equation A2 which yields:

$$
\begin{aligned}
\mathrm{Log}_{10}\ (2000^{\circ}\mathrm{F\ stress\ rupture\ life,\ hours}) = \\
& -7.088 + 3.68 + 9.54 + 4.06 + 0.146 \\
& \quad - 1.12 - 1.24 - 0.191 - 0.016 \\
& \quad\quad - 3.56 - 1.10 - 0.101 \\
& \quad\quad\quad - 1.38 - 0.015 \\
& \quad\quad\quad\quad - 0.0076 \\
& = -7.088 + 2.56 + 4.74 + 1.39 + 0.006 \\
& = 1.61
\end{aligned}
$$

or 2000°F stress rupture life, hours = 40.7.

TABLE A1

2000°F Stress Rupture Life for Series I

(Life in hours converted to Log_{10} (hours x 10^2) for computation)

A. 4.5% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	0.0000 0.0000 0.0000	(1.0)*	1.4771 1.3802 1.2553	(4.5)	2.7243 2.7853 2.7559	(8.0)
5.5	1.3802 1.4771 1.3222	(4.5)	2.9191 2.9777 2.8808	(8.0)	1.3222 1.5051 1.3802	(1.0)
10.0	3.1461 3.2227 3.2672	(8.0)	1.9542 2.0414 2.0000	(1.0)	2.1461 2.1761 2.2553	(4.5)

* Numbers in parentheses are Tantalum contents.

Average Test Value

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	1.5351 (0.34)**	2.0984 (1.25)	2.1167 (1.31)
W	1.3753 (0.24)	1.9072 (0.81)	2.4677 (2.94)
Ta	1.1337 (0.14)	1.6522 (0.45)	2.9643 (9.21)

** Average Life in hours is given in parenthesis.

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio***
Mo	1.9682	2	0.9841	28.52
W	5.3706	2	2.6853	77.83
Ta	16.0258	2	8.0129	232.26
Residual and Interaction	0.6907	20	0.0345	
Total	24.0553	26		

*** F ratio for 99% significance level $\geq$ 5.85

TABLE A1 (continued)

2000°F Stress Rupture Life for Series I

(Life in hours converted to $\log_{10}$ (hours x 10^2) for computation)

B. 6.3% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		Molybdenum 4.5		Molybdenum 8.0	
1.0	0.8451 0.7782 0.9031	(1.0)*	2.6721 2.5911 2.5682	(4.5)	2.3979 2.3617 2.4472	(8.0)
5.5	2.9736 3.0934 3.1399	(4.5)	2.6721 2.6232 2.5441	(8.0)	2.1139 2.1761 2.0792	(1.0)
10.0	2.7324 2.3424 2.7924	(8.0)	2.3424 2.3424 2.4472	(1.0)	1.5441 1.5441 1.6232	(4.5)

* Numbers in parentheses are Tantalum contents.

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	2.1778 (1.51)**	2.5336 (3.42)	2.0319 (1.08)
W	1.9516 (0.90)	2.6017 (4.00)	2.1901 (1.55)
Ta	1.7808 (0.60)	2.4166 (2.61)	2.5459 (3.52)

** Average Life in hours is given in parenthesis.

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio***
Mo	1.1988	2	0.5994	2.83
W	1.9468	2	0.9734	4.59
Ta	3.5023	2	1.7512	8.26
Residual & Interaction	4.2399	20	0.2120	
Total	10.8878	26		

*** F Ratio for 99% significance level ≥ 5.85

TABLE A1 (continued)

2000°F Stress Rupture Life for Series I

(Life in hours converted to $\log_{10}$ (hours x 10^2) for computation)

C. 6.3% Al - 1.8 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	1.0000 1.1139 1.0000	(1.0)*	2.3222 2.3979 2.4314	(4.5)	2.3010 2.3979 2.3617	(8.0)
5.5	2.7482 3.1106 2.7634	(4.5)	2.4771 2.3424 2.3617		2.4314 2.4150 2.4150	(1.0)
10.0	2.3010 2.4150 1.7782	(8.0)	2.7243 2.8921 2.6902	(1.0)	1.3979 1.3802 1.3010	(4.5)

* Numbers in parentheses are Tantalum contents.

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	2.0256 (1.06)**	2.5155 (3.28)	2.0446 (1.11)
W	1.9251 (0.84)	2.5628 (3.65)	2.0978 (1.25)
Ta	2.0758 (1.19)	2.2059 (1.61)	2.3040 (2.01)

** Average Life in hours is given in parenthesis.

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio***
Mo	1.3864	2	0.6932	2.42
W	1.9578	2	0.9789	3.42
Ta	0.2359	2	0.1180	0.41
Residual & Interaction	5.7306	20	0.2865	
Total	9.3107	26		

*** F ratio for 99% significance level $\geq$ 5.85

TABLE A2

2000°F Stress Rupture Elongation (%)
for Series I

A. 4.5% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	73.1 72.6 73.5	(1.0)*	29.8 23.1 33.1	(4.5)	5.3 3.7 6.9	(8.0)
5.5	23.3 23.8 28.8	(4.5)	4.0 6.1 4.9	(8.0)	25.5 40.3 29.2	(1.0)
10.0	6.6 2.6 7.4	(8.0)	26.1 25.9 25.9	(1.0)	9.4 9.3 12.2	(4.5)

* Numbers in parentheses are Tantalum contents.

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	34.6	19.9	15.8
W	35.7	20.7	13.9
Ta	43.6	21.4	5.3

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	1773.27	2	886.64	23.64
W	2231.03	2	1115.52	29.74
Ta	6651.17	2	3325.59	88.66
Residual & Interaction	750.16	20	37.51	
Total	11405.63	26		

** F Ratio for 99% significance level ≥ 5.85

TABLE A2 (continued)

2000°F Stress Rupture Elongation (%)
for Series I

B. 6.3% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	41.4 40.3 38.6	(1.0)*	1.4 4.2 7.6	(4.5)	2.7 4.1 5.1	(8.0)
5.5	9.7 7.9 8.1	(4.5)	4.4 2.3 2.5	(8.0)	9.1 9.0 10.3	(1.0)
10.0	2.5 3.3 2.8	(8.0)	6.7 5.9 7.4	(1.0)	14.9 16.3 14.8	(4.5)

* Numbers in parentheses are Tantalum contents

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	17.2	4.7	9.6
W	16.2	7.0	8.3
Ta	18.7	9.4	3.3

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	710.40	2	355.20	6.25
W	440.03	2	220.02	3.87
Ta	1088.54	2	544.27	9.58
Residual and Interaction	1135.85	20	56.79	
Total	3374.82	26		

** Ratio for 99% significance level ≥ 5.85

TABLE A2 (continued)

2000°F Stress Rupture Elongation (%) for Series I

C. 6.3% Al - 1.8 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	30.2 28.0 44.7	(1.0)*	16.9 11.8 17.2		4.2 4.7 4.2	(8.0)
5.5	3.5 3.3 1.8	(4.5)	4.7 3.8 5.4	(8.0)	6.4 5.6 7.1	(1.0)
10.0	3.4 7.0 6.5	(8.0)	7.0 4.5 7.6	(1.0)	12.6 10.4 14.6	(4.5)

* Numbers in parentheses are Tantalum contents

Average Test Value

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	14.3	8.8	7.8
W	18.0	4.6	8.2
Ta	15.7	10.2	4.9

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	221.00	2	110.50	2.34
W	862.69	2	431.35	9.13
Ta	524.88	2	262.44	5.56
Residual and Interaction	944.45	20		
Total	2553.02	26		

** F Ratio for 99% significance level $\geq$ 5.85

TABLE A3

Room Temp. Tensile Strength (10^3 psi) for
Series I

A. 4.5% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		Molybdenum 4.5		Molybdenum 8.0	
1.0	104.7 105.2	(1.0)*	120.6 125.3	(4.5)	142.1 143.9	(8.0)
5.5	128.1 141.8	(4.5)	143.9 142.7	(8.0)	116.7 116.7	(1.0)
10.0	148.2 148.2	(8.0)	125.9 124.5	(1.0)	152.4 152.2	(4.5)

* Number in parentheses are Tantalum contents.

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	129.4	130.5	137.3
W	123.6	131.7	141.9
Ta	115.6	136.7	144.8

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	223.28	2	111.64	6.21
W	1006.01	2	503.01	27.99
Ta	2730.28	2	1365.14	75.97
Residual & Interaction	197.70	11	17.97	
Total	4157.27	17		

** F Ratio for 99% significance level ≥ 7.20

TABLE A3 (Continued)

Room Temp. Tensile Strength (10^3 psi) for Series I i

C. 6.3% Al - 1.8 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	123.1 117.2	(1.0)*	134.1 137.1	(4.5)	148.6 147.3	(8.0)
5.5	137.8 144.4	(4.5)	137.6 138.6	(8.0)	156.2 152.4	(1.0)
10.0	135.4 145.2	(8.0)	146.7 150.2	(1.0)	89.7 94.0	(4.5)

* Number in parentheses are Tantalum contents

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	133.9	140.7	131.4
W	134.6	144.5	126.9
Ta	141.0	122.9	142.1

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	281.49	2	140.75	0.48
W	937.80	2	468.90	1.59
Ta	1401.49	2	700.75	2.37
Residual & Interaction	3250.74	11	295.52	
Total	5871.52	17		

** F Ratio for 99% significance level ≥ 7.20

TABLE A3 (Continued)

Room Temperature Tensile Strength (10^3 psi) for Series I

B. 6.3% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	106.3 95.9	(1.0)*	124.9 123.1	(4.5)	151.0 149.0	(8.0)
5.5	135.1 133.6	(4.5)	150.5 153.2	(8.0)	129.7 133.3	(1.0)
10.0	149.4 144.4	(8.0)	132.5 132.4	(1.0)	101.2 102.9	(4.5)

* Number in parentheses are Tantalum contents.

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	127.5	136.1	127.9
W	125.0	139.2	127.1
Ta	121.7	120.1	149.6

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	286.10	2	143.05	1.07
W	704.93	2	352.47	2.63
Ta	3296.23	2	1648.12	12.27
Residual & Interaction	1476.96	11	134.27	
Total	5764.22	17		

** F Ratio for 99% significance level ≥ 7.20

TABLE A4

Room Temp. Tensile Elongation (%) for Series I

A. 4.5% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		Molybdenum 4.5		Molybdenum 8.0	
1.0	4.1 5.6	(1.0)*	6.2 6.9	(4.5)	2.5 3.9	(8.0)
5.5	5.6 8.0	(4.5)	3.4 3.7	(8.0)	1.7 1.9	(1.0)
10.0	3.9 5.0	(8.0)	1.9 2.2	(1.0)	0.3 0.8	(4.5)

* Numbers in parentheses are Tantalum contents

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	5.4	4.1	1.9
W	4.9	4.1	2.4
Ta	2.9	4.6	3.7

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	37.89	2	18.95	18.05
W	19.79	2	9.90	9.43
Ta	9.02	2	4.51	4.30
Residual and Interaction	11.60	11	1.05	
Total	78.30	17		

** F Ratio for 99% significance level ≥ 7.20

TABLE A4 (Continued)

Room Temp. Tensile Elongation (%) for Series I

B. 6.3% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0	Molybdenum 4.5	Molybdenum 8.0
1.0	8.0 10.4 (1.0)*	5.1 6.5 (4.5)	1.2 1.0 (8.0)
5.5	7.1 6.8 (4.5)	0.6 0.4 (8.0)	1.4 1.8 (1.0)
10.0	0.2 0.3 (8.0)	2.1 1.8 (1.0)	0.0 0.0 (4.5)

* Numbers in parentheses are Tantalum contents

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	5.5	2.8	0.9
W	5.4	3.0	0.7
Ta	4.3	4.3	0.6

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	63.32	2	31.66	55.54
W	64.41	2	32.21	56.51
Ta	52.81	2	26.41	46.33
Residual and Interaction	6.24	11	0.57	
Total	186.78	17		

** F Ratio for 99% significance level ≥ 7.20

TABLE A4 (Continued)

Room Temp. Tensile Elongation (%) for Series I

C. 6.3% Al - 1.8 Ti Base

Tungsten	Molybdenum 1.0		Molybdenum 4.5		Molybdenum 8.0	
1.0	9.3 8.0	(1.0)*	7.7 9.3	(4.5)	0.0 0.6	(8.0)
5.5	4.6 10.1	(4.5)	0.0 0.3	(8.0)	0.5 0.8	(1.0)
10.0	0.3 0.2	(8.0)	0.7 1.4	(1.0)	1.6 0.6	(4.5)

* Numbers in parentheses are Tantalum contents

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	5.4	3.2	0.7
W	5.8	2.7	0.8
Ta	3.5	5.7	0.2

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	67.35	2	33.68	19.81
W	76.90	2	38.45	22.62
Ta	89.07	2	44.54	26.20
Residual & Interaction	18.74	11	1.70	
Total	252.06	17		

** F Ratio for 99% significance level ≥ 7.20

TABLE A5

1400°F Tensile Strength (10^3 psi) for
Series I

A. 4.5% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	92.0 88.1	(1.0)*	111.0 108.8	(4.5)	145.0 145.4	(8.0)
5.5	106.0 98.7	(4.5)	142.4 136.4	(8.0)	94.5 92.8	(1.0)
10.0	147.4 148.2	(8.0)	116.1 115.8	(1.0)	137.1 138.1	(4.5)

* Numbers in parentheses are Tantalum contents

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	113.4	121.8	125.5
W	115.1	111.8	133.8
Ta	99.9	116.6	144.1

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	459.34	2	229.67	17.14
W	1689.54	2	844.77	63.04
Ta	5990.47	2	2995.24	223.53
Residual & Interaction	147.43	11	13.40	
Total	8286.78	17		

** F Ratio for 99% significance level ≥ 7.20

TABLE A5 (Continued)

1400°F Tensile Strength (10^3 psi) for Series I

B. 6.3% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	84.5 99.8	(1.0)*	132.9 130.0	(4.5)	141.9 146.4	(8.0)
5.5	139.7 136.8	(4.5)	135.7 129.5	(8.0)	130.7 122.4	(1.0)
10.0	145.6 139.2	(8.0)	127.1 119.6	(1.0)	93.5 93.0	(4.5)

* Numbers in parentheses are Tantalum contents

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	124.3	129.1	121.3
W	122.6	132.5	119.7
Ta	114.0	121.0	139.7

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	186.98	2	93.49	0.30
W	540.06	2	270.03	0.86
Ta	2119.92	2	1059.96	3.37
Residual & Interaction	3464.29	11	314.94	
Total	6311.25	17		

** F Ratio for 99% significance level ≥ 7.20

TABLE A5 (Continued)

1400°F Tensile Strength (10^3 psi) for Series I

C. 6.3% Al - 1.8 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	116.3 114.0	(1.0)*	120.0 130.5	(4.5)	144.9 140.2	(8.0)
5.5	142.4 140.4	(4.5)	128.7 136.1	(8.0)	141.2 129.5	(1.0)
10.0	135.3 143.2	(8.0)	146.2 144.2	(1.0)	98.0 78.2	(4.5)

* Numbers in parentheses are Tantalum contents

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	131.9	134.3	122.0
W	127.7	136.4	124.2
Ta	131.9	118.3	138.1

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	510.15	2	255.08	0.81
W	473.26	2	236.63	0.76
Ta	1234.11	2	617.06	1.97
Residual & Interaction	3446.27	11	313.30	
Total	5663.79	17		

** F Ratio for 99% significance level $\geq$ 7.20

TABLE A6

1400°F Tensile Elongation (%)
for Series I

A. 4.5% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	2.9 1.8	(1.0)*	5.2 3.5	(4.5)	2.9 1.8	(8.0)
5.5	2.8 2.8	(4.5)	2.7 1.9	(8.0)	2.1 3.3	(1.0)
10.0	4.8 2.8	(8.0)	2.2 1.8	(1.0)	0.9 1.2	(4.5)

* Numbers in parentheses are Tantalum contents.

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	2.98	2.88	2.03
W	3.02	2.60	2.28
Ta	2.35	2.73	2.82

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	3.27	2	1.64	1.19
W	1.62	2	0.81	0.59
Ta	0.75	2	0.38	0.28
Residual & Interaction	15.22	11	1.38	
Total	20.86	17		

** F Ratio for 99% significance level ≥ 7.20

TABLE A6 (Continued)

1400°F Tensile Elongation (%) for Series I

B. 6.3% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	3.0 2.0	(1.0)*	3.9 3.3	(4.5)	0.8 1.1	(8.0)
5.5	4.2 3.1	(4.5)	0.4 0.6	(8.0)	1.2 1.4	(1.0)
10.0	0.6 1.0	(8.0)	0.4 0.4	(1.0)	0.7 0.6	(4.5)

* Numbers in parentheses are Tantalum contents.

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	2.32	1.50	0.97
W	2.35	1.82	0.62
Ta	1.40	2.63	0.75

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	5.54	2	2.77	13.19
W	9.46	2	4.73	22.52
Ta	11.56	2	5.78	27.52
Residual & Interaction	2.33	11	0.21	
Total	28.89	17		

** F Ratio for 99% significance level ≥ 7.20

TABLE A6 (Continued)

1400°F Tensile Elongation (%)
for Series I

C. 6.3% Al - 1.8 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	4.6 3.3	(1.0)*	3.0 5.7	(4.5)	0.9 0.3	(8.0)
5.5	2.6 3.0	(4.5)	0.2 0.2	(8.0)	1.3 0.9	(1.0)
10.0	0.3 0.9	(8.0)	0.3 0.8	(1.0)	0.6 0.5	(4.5)

* Numbers in parentheses are Tantalum contents.

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	2.45	1.70	0.75
W	2.97	1.37	0.57
Ta	1.87	2.57	0.47

	Sum of Squares	Degress of Freedom	Mean Square	F Ratio**
Mo	8.72	2	4.36	7.27
W	17.93	2	8.97	14.95
Ta	13.73	2	6.87	11.45
Residual & Interaction	6.62	11	0.60	
Total	47.00	17		

** F Ratio for 99% significance level ≥ 7.20

TABLE A7

1875°F Tensile Strength (10^3 psi) for Series I

A. 4.5% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	15.0 15.5	(1.0)*	29.2 29.5	(4.5)	44.3 46.7	(8.0)
5.5	29.5 29.7	(4.5)	50.1 51.0	(8.0)	36.2 37.0	(1.0)
10.0	54.1 56.0	(8.0)	42.1 43.9	(1.0)	44.3 41.7	(4.5)

* Numbers in parentheses are Tantalum contents.

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	33.3	41.0	41.7
W	30.0	38.9	47.0
Ta	31.6	34.0	50.4

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	259.75	2	129.88	22.05
W	865.91	2	432.96	73.51
Ta	1251.15	2	625.58	106.21
Residual & Interaction	64.77	11	5.89	
Total	2441.58	17		

** F Ratio for 99% significance level ≥ 7.20

TABLE A7 (Continued)

1875°F Tensile Strength (10^3 psi) for Series I

B. 6.3% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		Molybdenum 4.5		Molybdenum 8.0	
1.0	25.5 25.7	(1.0)*	46.5 47.7	(4.5)	43.7 46.6	(8.0)
5.5	51.7 49.9	(4.5)	49.2 48.5	(8.0)	43.2 42.9	(1.0)
10.0	51.2 50.0	(8.0)	44.5 45.2	(1.0)	36.4 34.8	(4.5)

* Numbers in parentheses are Tantalum contents.

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	42.3	46.9	41.3
W	39.3	47.6	43.7
Ta	37.8	44.5	48.2

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	108.83	2	54.42	1.40
W	206.11	2	103.06	2.64
Ta	331.21	2	165.61	4.25
Residual & Interaction	428.65	11	38.97	
Total	1074.80	17		

** F Ratio for 99% significance level $\geq$ 7.20

TABLE A7 (Continued)

1875°F Tensile Strength (10^3 psi) for Series I

C. 6.3% Al - 1.8 Ti Base

Tungsten	Molybdenum 1.0		Molybdenum 4.5		Molybdenum 8.0	
1.0	26.2 25.7	(1.0)*	39.4 39.5	(4.5)	45.5 49.5	(8.0)
5.5	56.9 55.1	(4.5)	45.3 44.5	(8.0)	46.3 43.5	(1.0)
10.0	46.4 47.2	(8.0)	45.9 47.5	(1.0)	34.7 33.2	(4.5)

* Numbers in parentheses are Tantalum contents.

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	42.9	43.7	42.1
W	37.6	48.6	42.5
Ta	39.2	43.1	46.4

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	7.36	2	3.68	0.06
W	362.41	2	181.21	2.81
Ta	156.71	2	78.36	1.21
Residual & Interaction	709.69	11	64.52	
Total	1236.17	17		

** F Ratio for 99% significance level ≥ 7.20

TABLE A8

1875°F Tensile Elongation (%)
for Series I

A. 4.5% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	31.0 22.7	(1.0)*	21.3 23.9	(4.5)	11.7 8.9	(8.0)
5.5	18.7 21.4	(4.5)	8.1 5.8	(8.0)	14.8 13.3	(1.0)
10.0	9.9 8.5	(8.0)	13.5 11.9	(1.0)	15.0 17.5	(4.5)

* Numbers in parentheses are Tantalum contents.

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	18.7	14.1	13.5
W	19.9	13.7	12.7
Ta	17.9	19.6	8.8

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	96.62	2	48.31	7.48
W	183.25	2	91.63	14.18
Ta	404.05	2	202.03	31.27
Residual & Interaction	71.10	11	6.46	
Total	755.02	17		

** F Ratio for 99% significance level $\geq$ 7.20

TABLE A8 (Continued)

1875°F Tensile Elongation (%)
for Series I

B. 6.3% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	- 22.3	(1.0)*	7.1 7.4	(4.5)	7.8 5.9	(8.0)
5.5	12.8 12.4	(4.5)	7.1 7.3	(8.0)	15.4 11.9	(1.0)
10.0	6.5 5.1	(8.0)	7.2 9.7	(1.0)	12.4 12.9	(4.5)

* Numbers in parentheses are Tantalum contents.

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	13.6	7.6	11.1
W	12.1	11.2	9.0
Ta	14.8	10.8	6.6

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	106.43	2	53.22	5.29
W	31.53	2	15.77	1.57
Ta	200.96	2	100.48	9.98
Residual & Interaction	110.78	11	10.07	
Total	449.70	17		

** F Ratio for 99% significance level $\geq$ 7.20

TABLE A8 (Continued)

1875°F Tensile Elongation (%)
for Series I

C. 6.3% Al - 1.8 Ti Base

Tungsten \ Molybdenum	1.0		4.5		8.0	
1.0	19.5 25.0	(1.0)*	16.1 16.5	(4.5)	6.8 7.3	(8.0)
5.5	3.8 4.5	(4.5)	7.9 7.7	(8.0)	11.7 10.2	(1.0)
10.0	5.1 4.0	(8.0)	9.4 8.9	(1.0)	10.3 9.2	(4.5)

* Numbers in parentheses are Tantalum contents.

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	10.3	11.1	9.3
W	15.2	7.6	7.8
Ta	14.1	10.1	6.5

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	10.17	2	5.09	0.38
W	223.60	2	111.80	8.45
Ta	175.77	2	87.89	6.64
Residual & Interaction	145.48	11	13.23	
Total	555.02	17		

** F Ratio for 99% significance level $\geq$ 7.20

TABLE A9

2000°F Tensile Strength (10^3 psi) for Series I

A. 4.5% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		Molybdenum 4.5		Molybdenum 8.0	
1.0	5.0 5.0	(1.0)*	14.3 14.3	(4.5)	28.3 30.4	(8.0)
5.5	16.0 15.6	(4.5)	30.5 29.5	(8.0)	17.8 20.0	(1.0)
10.0	36.1 34.7	(8.0)	23.0 23.1	(1.0)	26.4 25.2	(4.5)

* Numbers in parentheses are Tantalum contents.

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	18.7	22.5	24.7
W	16.2	21.6	28.1
Ta	15.7	18.6	31.6

	Sum of Sqaures	Degrees of Freedom	Mean Square	F Ratio**
Mo	110.07	2	55.04	19.52
W	423.81	2	211.91	75.15
Ta	860.95	2	430.48	152.65
Residual & Interaction	30.97	11	2.82	
Total	1425.80	17		

** F Ratio for 99% significance level ≥ 7.20

TABLE A9 (Continued)

2000°F Tensile Strength (10^3 psi) for
Series I

B. 6.3% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		Molybdenum 4.5		Molybdenum 8.0	
1.0	11.1 11.1	(1.0)*	28.3 28.0	(4.5)	30.6 30.5	(8.0)
5.5	32.5 31.3	(4.5)	30.6 32.3	(8.0)	24.8 26.0	(1.0)
10.0	32.7 32.6	(8.0)	26.1 24.7	(1.0)	21.5 21.4	(4.5)

* Numbers in parentheses are Tantalum contents.

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	25.2	28.3	25.8
W	23.3	29.6	26.5
Ta	20.6	27.2	31.6

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	32.94	2	16.47	0.72
W	119.72	2	59.86	2.62
Ta	362.15	2	181.08	7.92
Residual & Interaction	240.45	11	22.86	
Total	755.26	17		

** F Ratio for 99% significance level $\geq$ 7.20

TABLE A9 (Continued)

2000°F Tensile Strength (10^3 psi) for Series I

C. 6.3% Al - 1.8 Ti Base

Tungsten	Molybdenum 1.0		Molybdenum 4.5		Molybdenum 8.0	
1.0	11.3 11.9	(1.0)*	24.0 23.6	(4.5)	30.3 31.1	(8.0)
5.5	34.4 35.5	(4.5)	29.0 28.1	(8.0)	30.2 28.9	(1.0)
10.0	31.6 31.4	(8.0)	29.4 24.3	(1.0)	19.8 19.8	(4.5)

* Numbers in parentheses are Tantalum contents.

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	26.0	26.4	26.7
W	22.0	31.0	26.1
Ta	22.7	26.2	30.3

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	1.34	2	0.67	0.02
W	243.01	2	121.51	3.33
Ta	172.83	2	86.42	2.37
Residual & Interaction	401.68	11	36.52	
Total	818.86	17		

** F Ratio for 99% significance level $\geq$ 7.20

TABLE A10

2000°F Tensile Elongation (%)
for Series I

A. 4.5% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	72.6 71.8	(1.0)*	27.1 28.2	(4.5)	13.1 13.8	(8.0)
5.5	30.3 26.4	(4.5)	11.3 10.5	(8.0)	26.5 20.6	(1.0)
10.0	10.0 15.7	(8.0)	22.2 16.5	(1.0)	15.5 10.9	(4.5)

* Numbers in parentheses are Tantalum contents.

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	37.8	19.3	16.7
W	37.8	20.9	15.1
Ta	38.4	23.1	12.4

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	1585.29	2	792.65	15.11
W	1658.55	2	829.28	15.81
Ta	2044.28	2	1022.14	19.48
Residual & Interaction	577.10	11	52.46	
Total	5865.22	17		

** F Ratio for 99% significance level ≥ 7.20

TABLE A10 (Continued)

2000°F Tensile Elongation (%) for Series I

B. 6.3% Al - 1.0 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	42.1 32.5	(1.0)*	6.0 12.0	(4.5)	8.1 10.8	(8.0)
5.5	17.5 13.2	(4.5)	10.6 8.6	(8.0)	15.6 11.3	(1.0)
10.0	5.5 7.2	(8.0)	10.7 16.2	(1.0)	16.4 15.1	(4.5)

* Numbers in parentheses are Tantalum contents.

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	19.7	10.7	12.9
W	18.6	12.8	11.9
Ta	21.4	13.4	8.5

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	263.11	2	131.56	2.82
W	159.38	2	79.69	1.71
Ta	511.64	2	255.82	5.49
Residual & Interaction	512.39	11	46.58	
Total	1446.52	17		

** F Ratio for 99% significance level $\geq$ 7.20

TABLE A10 (Continued)

2000°F Tensile Elongation (%)
for Series I

C. 6.3% Al - 1.8 Ti Base

Tungsten	Molybdenum 1.0		4.5		8.0	
1.0	35.6 42.9	(1.0)*	15.4 22.8	(4.5)	8.6 10.7	(8.0)
5.5	5.2 3.7	(4.5)	5.7 8.3	(8.0)	13.0 12.3	(1.0)
10.0	10.7 6.9	(8.0)	13.2 11.4	(1.0)	17.3 18.5	(4.5)

* Numbers in parentheses are Tantalum contents.

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	17.5	12.8	13.4
W	22.7	8.0	13.0
Ta	21.4	13.8	8.5

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio**
Mo	78.52	2	39.26	0.78
W	664.50	2	332.25	6.58
Ta	505.58	2	252.79	5.01
Residual & Interaction	555.48	11	50.50	
Total	1804.08	17		

** F Ratio for 99% significance level ≥ 7.20

TABIE A11

2000°F Stress Rupture Life for Series II
(Life in hours converted to $\log_{10}$(hours) for computation)

Vanadium	Columbium 0	Columbium 1.0	Columbium 2.0
0	0.8716 0.9956 (0) * 0.8129	1.2718 1.2380 (1.0) 1.1875	1.1700 1.4232 (2.0) 1.2146
1.0	1.1767 1.1605 (1.0) 1.1268	1.1072 1.1225 (2.0) 1.1732	1.0212 0.9731 (0) 0.8122
2.0	1.1000 1.1239 (2.0) 1.1332	0.8195 0.6684 (0) 0.5798	0.8476 0.8808 (1.0) 0.8014

* Numbers in parenthesis are Hafnium contents.

Average Test Value

	0%	1.0%	2.0%
Cb	1.0557(11.4)**	1.0186(10.4)	1.0160(10.4)
V	1.1317(13.5)	1.0748(11.9)	0.8838(7.6)
Hf	0.8394(6.9)	1.0768(11.9)	1.1742(14.9)

** Average Life in hours is given in parenthesis.

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio***
Cb	0.0088	2	0.0044	0.43
V	0.3034	2	0.1517	14.75
Hf	0.5339	2	0.2670	25.96
Residuals & Interactions	0.2057	20	0.0103	
Total	1.0518	26		

*** F Ratio for 99% significance level ≥ 5.85%

TABLE A12

2000°F Stress Rupture Elongation (%)
for Series II

Vanadium	Columbium 0	Columbium 1.0	Columbium 2.0
0	7.8 3.8 (0) * 6.9	13.1 9.9 (1.0) 13.5	4.7 9.2 (2.0) 2.9
1.0	19.0 12.1 (1.0) 18.6	11.2 7.7 (2.0) 5.0	2.5 1.2 (0) 3.10
2.0	7.7 6.0 (2.0) 9.1	2.2 4.9 (0) 3.4	9.1 9.8 (1.0) 7.9

* Numbers in parenthesis are Hafnium contents.

Average Test Value

	0%	1.0%	2.0%
Cb	10.1	7.9	5.6
V	8.0	8.9	6.7
Hf	4.0	12.6	7.1

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio **
Cb	91.58	2	45.79	8.62
V	23.07	2	11.54	2.17
Hf	339.90	2	169.95	32.00
Residual & Interactions	106.21	20	5.31	
Total	560.76	26		

** F Ratio for 99% significance level $\geq$ 5.85.

TABLE A13

Room Temp. Tensile Strength (10^3 psi) for Series II

Vanadium	Columbium 0	Columbium 1.0	Columbium 2.0
0	145.9, 147.0 (0)*	151.0, 151.4 (1.0)	147.3, 152.2 (2.0)
1.0	155.0, 151.6 (1.0)	153.6, 149.0 (2.0)	167.3, 164.2 (0)
2.0	149.2, 154.6 (2.0)	161.0, 151.8 (0)	153.8, 151.6 (1.0)

* Numbers in parenthesis are Hafnium contents

Average Test Value

	0%	1.0%	2.0%
Cb	150.6	153.0	156.1
V	149.1	156.8	153.7
Hf	156.2	152.4	151.0

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio **
Cb	91.77	2	45.88	2.45
V	177.58	2	88.79	4.74
Hf	87.33	2	43.67	2.33
Residual & Interactions	206.04	11	18.73	
Total	562.72	17		

** F Ratio for 99% significance level $\geq$ 7.20.

TABLE A14

Room Temp. Tensile Elongation (%) for Series II

Vanadium	Columbium 0	Columbium 1.0	Columbium 2.0
0	6.2, 6.6 (0)*	2.8, 3.7 (1.0)	1.6, 1.4 (2.0)
1.0	5.6, 6.0 (1.0)	0.5, 1.7 (2.0)	6.4, 5.0 (0)
2.0	3.2, 4.6 (2.0)	5.3, 3.7 (0)	1.8, 0.7 (1.0)

* Numbers in parenthesis are Hafnium contents

Average Test Value

	0%	1.0%	2.0%
Cb	5.4	3.0	2.8
V	3.7	4.2	3.2
Hf	5.5	3.4	2.2

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio **
Cb	24.72	2	12.36	12.14
V	2.90	2	1.45	1.42
Hf	34.70	2	17.35	17.04
Residual & Interactions	11.20	11	1.02	
Total	73.52	17		

** F Ratio for 99% significance level $\geq$ 7.20.

TABLE A15

1400°F Tensile Strength (10^3 psi) for Series II

Vanadium	Columbium 0	Columbium 1.0	Columbium 2.0
0	135.2, 133.5 (0)*	150.0, 150.2 (1.0)	158.0, 159.5 (2.0)
1.0	147.8, 147.0 (1.0)	154.0, 157.2 (2.0)	156.6, 154.2 (0)
2.0	152.4, 156.8 (2.0)	152.2, 152.2 (0)	156.6, 160.2 (1.0)

* Numbers in parenthesis are Hafnium contents

Average Test Values

	0%	1.0%	2.0%
Cb	145.4	152.6	157.5
V	147.7	152.8	155.1
Hf	147.3	152.0	156.3

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio **
Cb	442.1	2	221.0	30.14
V	169.2	2	84.6	11.53
Hf	243.1	2	121.5	16.57
Residual & Interactions	80.6	11	7.3	
Total	935.0	17		

** F Ratio for 99% significance level ≥ 7.20

TABLE A16

1400°F Tensile Elongation (%) for Series II

Vanadium	Columbium 0	Columbium 1.0	Columbium 2.0
0	2.8, 3.3 (0)*	4.5, 4.8 (1.0)	3.2, 4.1 (2.0)
1.0	5.1, 5.0 (1.0)	5.0, 5.9 (2.0)	3.8, 3.4 (0)
2.0	5.0, 3.0 (2.0)	2.2, 1.8 (0)	2.2, 2.0 (1.0)

* Numbers in parentheses are Hafnium contents

Average Test Values

	0%	1.0%	2.0%
Cb	4.0	4.0	3.1
V	3.8	4.7	2.7
Hf	2.9	3.9	4.4

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio **
Cb	3.36	2	1.68	4.29
V	12.03	2	6.01	15.36
Hf	6.98	2	3.49	8.92
Residual & Interactions	4.31	11	0.39	
Total	26.68	17		

** F Ratio for 99% significance level ≥ 7.20.

TABLE A17

1875°F Tensile Strength (10^3 psi) for Series II

Vanadium	Columbium 0	Columbium 1.0	Columbium 2.0
0	48.7, 50.3 (0)*	54.9, 53.0 (1.0)	59.6, 56.9 (2.0)
1.0	46.4, 45.8 (1.0)	54.9, 53.8 (2.0)	53.1, 53.7 (0)
2.0	54.5, 53.0 (2.0)	49.2, 47.8 (0)	45.3, 45.7 (1.0)

* Numbers in parenthesis are Hafnium contents

Average Test Value

	0%	1.0%	2.0%
Cb	49.8	52.3	52.4
V	53.9	51.3	49.2
Hf	50.5	48.5	55.4

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio **
Cb	25.88	2	12.94	2.44
V	65.21	2	32.60	6.14
Hf	153.42	2	76.71	14.45
Residual & Interactions	58.40	11	5.31	
Total	302.91	17		

** F Ratio for 99% significance level ≥ 7.20.

TABLE A18

1875°F Tensile Elongation (%) for Series II

Vanadium	Columbium 0	Columbium 1.0	Columbium 2.0
0	13.3, 6.5 (0)*	13.8, 14.4 (1.0)	10.8, 7.1 (2.0)
1.0	12.7, 14.8 (1.0)	8.0, 11.1 (2.0)	8.3, 8.2 (0)
2.0	12.1, 15.2 (2.0)	6.2, 6.4 (0)	4.9, 8.4 (1.0)

* Numbers in parenthesis are Hafnium contents

Average Test Value

	0%	1.0%	2.0%
Cb	12.4	10.0	8.0
V	11.0	10.5	8.9
Hf	8.2	11.5	10.7

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio **
Cb	60.47	2	30.24	4.02
V	14.84	2	7.42	1.00
Hf	36.85	2	18.42	2.45
Residual & Interactions	82.65	11	7.51	
Total	194.81	17		

** F Ratio for 99% significance level $\geq$ 7.20.

TABLE A19

2000°F Tensile Strength (10^3 psi) for Series II

Vanadium	Columbium 0	Columbium 1.0	Columbium 2.0
0	31.1, 31.7 (0)*	34.6, 34.6 (1.0)	37.9, 39.4 (2.0)
1.0	33.1, 31.7 (1.0)	36.8, 34.8 (2.0)	35.4, 37.0 (0)
2.0	33.3, 33.4 (2.0)	31.9, 31.5 (0)	31.3, 31.8 (1.0)

* Numbers in parenthesis are Hafnium contents

Average Test Value

	0%	1.0%	2.0%
Cb	32.4	34.0	35.5
V	34.9	34.8	32.2
Hf	33.1	32.8	35.9

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio **
Cb	28.57	2	14.28	9.67
V	27.94	2	13.97	9.46
Hf	35.20	2	17.60	11.92
Residual & Interactions	16.23	11	1.48	
Total	107.94	17		

** F Ratio for 99% significance level $\geq$ 7.20.

TABLE A20

2000°F Tensile Elongation (%) for Series II

Vanadium	Columbium 0	Columbium 1.0	Columbium 2.0
0	8.6 9.9 (0)*	13.8 17.3 (1.0)	9.7 7.7 (2.0)
1.0	17.6 15.7 (1.0)	7.7 8.7 (2.0)	9.1 7.7 (0)
2.0	15.5 11.9 (2.0)	8.9 12.6 (0)	10.7 9.6 (1.0)

* Numbers in parenthesis are Hafnium contents

Average Test Value

	0%	1.0%	2.0%
Cb	13.2	11.5	9.1
V	11.2	11.1	11.5
Hf	9.5	14.1	10.2

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio **
Cb	51.35	2	25.68	4.33
V	0.68	2	0.34	0.06
Hf	75.00	2	37.50	6.32
Residual & Interactions	65.23	11	5.93	
Total	192.26	17		

** F Ratio for 99% significance level $\geq$ 7.20.

TABLE A21

2000°F Stress Rupture Life for Series III

(Life in hours converted to $\log_{10}$(hours) for computation)

Ruthenium	Rhenium 0	2.0	4.0
0	0.8797 0.8222 (6.3) * 0.8808	1.1793 1.2770 (5.4) 1.2014	1.1903 1.2304 (4.5) 1.1485
1.0	0.9868 0.9085 (5.4) 0.9201	1.0000 1.0382 (4.5) 1.1139	0.5911 0.6180 (6.3) 0.6435
2.0	0.9096 0.9542 (4.5) 0.9350	0.7292 0.6415 (6.3) 0.7243	1.1004 1.0569 (5.4) 1.0086

* Numbers in parenthesis are Aluminum contents.

Average Test Value

	0% or 4.5%	1.0% or 2.0% or 5.4%	2.0% or 4.0% or 6.3%
Re	0.9108 (8.1) **	0.9894 (9.8)	0.9542 (9.0)
Ru	1.0900 (12.3)	0.8689 (7.4)	0.8955 (7.9)
Al	1.0578(11.4)	1.0710 (11.8)	0.7256 (5.3)

** Average life in hours is given in parenthesis.

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio ***
Re	0.0279	2	0.0140	5.25
Ru	0.2621	2	0.1311	49.28
Al	0.6895	2	0.3448	129.62
Residual & Interactions	0.0533	20	0.0026	
Total	1.0328	26		

*** F Ratio for 99% significant level $\geq$ 5.85

TABLE A22

2000°F Stress Rupture Elongation (%)
for Series III

Ruthenium	Rhenium 0	Rhenium 2.0	Rhenium 4.0
0	3.3 2.3 (6.3) * 1.8	5.0 1.6 (5.4) 0.8	2.1 2.4 (4.5) 3.3
1.0	1.8 1.3 (5.4) 2.5	2.4 2.6 (4.5) 3.9	2.2 2.3 (6.3) 2.0
2.0	3.7 3.2 (4.5) 1.8	2.1 - (6.3) 2.4	1.7 0.8 (5.4) 1.6

* Numbers in parenthesis are Aluminum contents.

Average Test Value

	0% or 4.5%	1.0% or 2.0% or 5.4%	2.0% or 4.0% or 6.3%
Re	2.4	2.6	2.0
Ru	2.5	2.3	2.2
Al	2.8	1.9	2.3

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio **
Re	1.27	2	0.64	0.75
Ru	0.52	2	0.26	0.31
Al	3.85	2	1.93	2.28
Residual & Interactions	16.87	20	0.84	
Total	22.51	26		

** F Ratio for 99% significance level $\geq$ 5.85.

TABLE A23

Room Temp. Tensile Strength (10^3 psi) for Series III

Ruthenium	Rhenium 0	Rhenium 2.0	Rhenium 4.0
0	145.5 147.1 (6.3)*	155.4 155.0 (5.4)	149.4 161.4 (4.5)
1.0	157.0 149.0 (5.4)	157.6 158.4 (4.5)	145.6 150.4 (6.3)
2.0	151.2 150.8 (4.5)	153.0 152.2 (6.3)	162.4 165.9 (5.4)

* Numbers in parenthesis are Aluminum contents.

Average Test Value

	0% or 4.5%	1.0% or 2.0% or 5.4%	2.0% or 4.0% or 6.3%
Re	150.1	155.3	155.8
Ru	152.3	153.0	155.9
Al	154.8	157.4	149.0

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio **
Re	120.20	2	60.10	3.45
Ru	44.15	2	22.08	1.27
Al	226.03	2	113.02	6.49
Residual & Interactions	191.47	11	17.41	
Total	581.85	17		

** F Ratio for 99% significance level ≥ 7.20.

TABLE A24

Room Temp. Tensile Elongation (%) for Series III

Ruthenium	Rhenium 0	Rhenium 2.0	Rhenium 4.0
0	4.1, 3.6 (6.3)*	4.6, 5.7 (5.4)	4.8, 6.5 (4.5)
1.0	4.4, 4.8 (5.4)	6.0, 6.5 (4.5)	0.2, 0.2 (6.3)
2.0	6.7, 6.5 (4.5)	1.2, 1.6 (6.3)	1.7, 2.8 (5.4)

* Numbers in parenthesis are Aluminum contents

Average Test Value

	0% or 4.5%	1.0% or 2.0% or 5.4%	2.0% or 4.0% or 6.3%
Re	5.0	4.3	2.7
Ru	4.9	3.7	3.4
Al	6.2	4.0	1.8

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio **
Re	16.87	2	8.44	28.54
Ru	7.35	2	3.67	12.43
Al	56.94	2	28.47	96.32
Residual & Interactions	3.25	11	0.30	
Total	84.41	17		

** F Ratio for 99% significance level ≥ 7.20 .

TABLE A25

1400°F Tensile Strength (10^3 psi) for Series III

Ruthenium	Rhenium 0	Rhenium 2.0	Rhenium 4.0
0	151.4 150.6 (6.3)*	159.6 154.6 (5.4)	151.2 148.4 (4.5)
1.0	152.2 151.8 (5.4)	149.2 144.4 (4.5)	139.5 122.1 (6.3)
2.0	139.8 142.0 (4.5	140.1 144.9 (6.3)	157.6 160.6 (5.4)

* Numbers in parenthesis are Aluminum Contents

Average Test Value

	0% or 4.5%	1.0% or 2.0% or 5.4%	2.0% or 4.0% or 6.3%
Re	148.0	148.8	146.6
Ru	152.6	143.2	147.5
Al	145.8	156.1	141.4

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio **
Re	15.27	2	7.63	0.18
Ru	267.65	2	133.82	3.18
Al	676.43	2	338.22	8.05
Residual & Interactions	468.33	11	42.03	
Total	1421.68	17		

** F Ratio for 99% significance level $\geq$ 7.20.

TABLE A26

1400°F Tensile Elongation (%) for
Series III

Ruthenium	Rhenium		
	0	2.0	4.0
0	2.2 3.3 (6.3)*	3.3 2.5 (5.4)	4.3 3.0 (4.5)
1.0	2.9 3.9 (5.4)	3.2 2.5 (4.5)	0.6 0.1 (6.3)
2.0	3.7 4.3 (4.5)	1.2 1.3 (6.3)	1.5 1.1 (5.4)

* Numbers in parenthesis are Aluminum contents

Average Test Value

	0% or 4.5%		1.0% or 2.0% or 5.4%			2.0% or 4.0% or 6.3%		
	0%	4.5%	1.0%	2.0%	5.4%	2.0%	4.0%	6.3%
Re	3.4			2.3			1.8	
Ru	3.1		2.2			2.2		
Al		3.5			2.5			1.5

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio **
Re	8.07	2	4.04	12.29
Ru	3.30	2	1.65	5.02
Al	12.62	2	6.31	19.21
Residual & Interactions	3.62	11	0.33	
Total	27.61	17		

** F Ratio for 99% significance level ≥ 7.20.

TABLE A27

1875°F Tensile Strength (10^3 psi)
for Series III

Ruthenium	Rhenium 0	Rhenium 2.0	Rhenium 4.0
0	54.7, 53.6 (6.3)*	64.8, 65.5 (5.4)	63.9, 60.4 (4.5)
1.0	53.1, 52.1 (5.4)	55.3, 53.4 (4.5)	57.7, 57.3 (6.3)
2.0	57.8, 56.2 (4.5)	55.2, 54.9 (6.3)	73.5, 74.0 (5.4)

* Numbers in parenthesis are Aluminum contents

Average Test Value

	0%	or 4.5%	1.0%	or 2.0%	or 5.4%	2.0%	or 4.0%	or 6.3%
Re	54.6			58.2			64.5	
Ru	60.5		54.8			61.9		
Al		57.8			63.8			55.6

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio **
Re	300.24	2	150.12	24.17
Ru	169.72	2	84.86	13.66
Al	218.96	2	109.48	17.63
Residual & Interactions	68.32	11	6.21	
Total	757.24	17		

** F Ratio for 99% significance level ≥ 7.20.

TABLE A28

<u>1875°F Tensile Elongation (%) for Series III</u>

Ruthenium	Rhenium 0	Rhenium 2.0	Rhenium 4.0
0	5.7 3.4 (6.3)*	4.9 5.1 (5.4)	6.8 8.6 (4.5)
1.0	2.5 3.3 (5.4)	6.0 7.0 (4.5)	2.5 2.2 (6.3)
2.0	7.2 5.8 (4.5)	3.7 4.1 (6.3)	2.5 3.1 (5.4)

* Numbers in parenthesis are Aluminum contents

Average Test Value

	0% or 4.5%	1.0% or 2.0% or 5.4%	2.0% or 4.0% or 6.3%
Re	4.6	5.1	4.3
Ru	5.8	3.9	4.4
Al	6.9	3.6	3.6

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio **
Re	2.18	2	1.09	1.82
Ru	10.83	2	5.42	9.06
Al	44.00	2	22.00	36.79
Residual & Interactions	6.59	11	0.60	
Total	63.60	17		

** F Ratio for 99% significance level $\geq$ 7.20.

TABLE A29

2000°F Tensile Strength (10^3 psi)
for Series III

Ruthenium	Rhenium 0	Rhenium 2.0	Rhenium 4.0
0	41.1, 39.9 (6.3)*	39.3, 43.0 (5.4)	38.6, 38.0 (4.5)
1.0	39.0, 39.0 (5.4)	39.3, 36.4 (4.5)	37.4, 38.6 (6.3)
2.0	33.0, 32.5 (4.5)	36.7, 36.5 (6.3)	46.7, 47.3 (5.4)

* Numbers in parenthesis are Aluminum contents

Average Test Value

	0% or 4.5%	1.0% or 2.0% or 5.4%	2.0% or 4.0% or 6.3%
Re	37.4	38.5	41.1
Ru	40.0	38.3	38.8
Al	36.3	42.4	38.4

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio **
Re	42.80	2	21.40	2.83
Ru	9.16	2	4.58	0.60
Al	114.82	2	57.41	7.59
Residual & Interactions	83.22	11	7.56	
Total	250.00	17		

** F Ratio for 99% significance level $\geq$ 7.20.

TABLE A30

2000°F Tensile Elongation (%) for Series III

Ruthenium	Rhenium 0	Rhenium 2.0	Rhenium 4.0
0	5.0, 5.6 (6.3)*	6.0, 6.1 (5.4)	6.8, 9.4 (4.5)
1.0	3.4, 2.9 (5.4)	6.4, 7.2 (4.5)	6.0, 5.3 (6.3)
2.0	10.4, 9.5 (4.5)	5.5, 6.1 (6.3)	5.2, 4.7 (5.4)

* Numbers in parenthesis are Aluminum contents

Average Test Value

	0%	or 4.5%	1.0%	or 2.0%	or 5.4%	2.0%	or 4.0%	or 6.3%
Re	6.1			6.2			6.2	
Ru	6.5		5.2			6.9		
Al		8.3			4.7			5.6

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio **
Re	0.03	2	0.02	0.01
Ru	9.42	2	4.71	3.61
Al	41.52	2	20.76	15.89
Residual & Interactions	14.38	11	1.31	
Total	65.35	17		

** F Ratio for 99% significance level $\geq$ 7.20.

TABLE A31

Workability Results for Series I

A. 4.5% Al - 1.0 Ti Base

Tungsten	Molybdenum		
	1.0	4.5	8.0
1.0	2.5(1.0)*	2.5(4.5)	3.0(8.0)
5.5	3.0(4.5)	3.0(8.0)	3.0(1.0)
10.0	4.0(8.0)	3.0(1.0)	3.5(4.5)

* Numbers in parenthesis are Tantalum contents

Average Test Values

	1%	4.5% or 5.5%		8.0% or 10.0%	
Mo	3.17	2.83		3.17	
W	2.67		3.00		3.50
Ta	2.83	3.00		3.33	

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio **
Mo	0.444	2	0.222	0.89
W	2.111	2	1.056	4.24
Ta	0.778	2	0.389	1.56
Residual & Interactions	2.736	29	0.249	
Total	6.069	35		

** F Ratio for 99% significance level ≥ 5.52.

TABLE A31 (Continued)

Workability Results for Series I

B. 6.3% Al - 1.0 Ti Base

Tungsten	Molybdenum		
	1.0	4.5	8.0
1.0	2.0(1.0)*	3.0(4.5)	4.0(8.0)
5.5	3.5(4.5)	4.0(8.0)	3.0(1.0)
10.0	4.0(8.0)	3.5(1.0)	4.0(4.5)

* Number in parenthesis are Tantalum contents

Average Test Values

	1.0%	4.5%	5.5%	8.0%	10.0%
Mo	3.17	3.50		3.67	
W	3.00		3.50		3.83
Ta	2.83	3.50		4.00	

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio **
Mo	0.778	2	0.389	3.24
W	2.111	2	1.056	8.80
Ta	4.111	2	2.056	17.14
Residual & Interactions	1.319	29	0.120	
Total	8.319	35		

** F Ratio for 99% significance level ≥ 5.52.

TABLE A31 (Continued)

Workability Results for Series I

C. 6.3% Al - 1.8 Ti Base

Tungsten	Molybdenum		
	1.0	4.5	8.0
1.0	2.0(1.0)*	2.5(4.5)	3.5(8.0)
5.5	3.5(4.5)	5.0(8.0)	3.5(1.0)
10.0	4.0(8.0)	4.0(1.0)	5.0(4.5)

* Numbers in parenthesis are Tantalum contents

Average Test Values

	1%	4.5% or 5.5%	8.0% or 10.0%
Mo	3.17	3.83	4.00
W	2.67	4.00	4.33
Ta	3.17	3.67	4.17

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio **
Mo	2.333	2	1.167	4.34
W	9.333	2	4.667	17.35
Ta	3.000	2	1.500	5.58
Residual & Interactions	2.959	29	0.269	
Total	17.625	35		

** F Ratio for 99% significance level ≥ 5.52.

TABLE A32

Workability Results for Series II

Vanadium	Columbium		
	0	1.0	2.0
0	2.5(0)*	3.5(1.0)	5.0(2.0)
1.0	3.0(1.0)	4.0(2.0)	3.5(0)
2.0	3.5(2.0)	4.5(0)	5.0(1.0)

* Numbers in parenthesis are Hafnium contents

	Average Test Value		
	0%	1.0%	2.0%
Cb	3.00	4.00	4.50
V	3.67	3.50	4.33
Hf	3.50	3.83	4.17

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio **
Cb	7.000	2	3.500	13.59
V	2.333	2	1.167	4.53
Hf	1.333	2	0.667	2.59
Residual & Interactions	2.834	29	0.258	
Total	13.500	35		

** F Ratio for 99% significance level ≥ 5.52.

TABLE A33

Workability Results for Series III

Ruthenium	Rhenium 0	Rhenium 2.0	Rhenium 4.0
0	3.5(6.3)*	3.5(5.4)	3.0(4.5)
1.0	3.0(5.4)	3.0(4.5)	5.0(6.3)
2.0	3.0(4.5)	4.0(6.3)	6.0(5.4)

* Numbers in parenthesis are Aluminum contents

Average Test Value

	0% or 4.5%	1.0% or 2.0% or 5.4%	2.0% or 4.0% or 6.3%
Re	3.17	3.50	4.67
Ru	3.33	3.67	4.33
Al	3.0	4.17	4.17

	Sum of Squares	Degrees of Freedom	Mean Square	F Ratio **
Re	7.444	2	3.722	11.34
Ru	3.111	2	1.556	4.74
Al	5.444	2	2.722	8.29
Residual & Interactions	3.612	29	0.328	
Total	19.611	35		

** F Ratio for 99% significance level ≥ 5.52.

TABLE B1

Yates' Method of Analysis Using % Elongation Data of 2000°F Stress Rupture Tests from Series IV

Treatment Combination	Response (Ave.% Elong.)	3	4	5	6	2000°F Stress Rupture Elong. %	Estimated Effects **
gh (L)*	3.97	6.24	19.24	33.54	69.52	8.688	Total
ae (A)	2.27	13.00	14.30	35.98	4.44	0.554	Co
befh (C)	5.73	8.27	18.24	1.28	10.94	1.371	Cr
abfg (O)	7.27	6.03	17.74	3.16	2.70	0.338	CoCr+TaW-HfC-ReZr
cefg (M)	3.60	6.94	-0.16	4.52	-5.44	-0.679	Ta
acfh (D)	4.67	11.30	1.44	6.42	-1.56	-0.196	CoTa+CrW-ReC-HfZr
bc (G)	2.83	7.84	3.16	2.54	-11.30	-1.412	CrTa+CoW+HfRe+CZr
abcegh (P)	3.20	9.90	0	0.16	-2.10	-0.262	W
df (N)	2.47	-1.70	6.76	-4.94	2.44	0.304	Hf
adefgh (J)	4.47	1.54	-2.24	-0.50	1.88	0.238	CoHf+WRe-CrC-TaZr
bdeg (F)	5.07	1.07	4.36	1.60	1.90	0.238	CrHf-CoC+TaRe-WZr
abdh (K)	6.23	0.37	2.06	-3.16	-2.38	-0.296	-C
cdeh (I)	4.17	2.00	3.24	-9.00	4.44	0.554	TaHf-CoZr+CrRe-WC
acdg (H)	3.67	1.16	-0.70	-2.30	-4.76	-0.596	-Zr
bcdfgh (E)	4.70	-0.50	-0.84	-3.94	6.70	0.838	Re
abcdef (B)	5.20	0.50	1.00	1.84	5.78	0.721	CoRe+HfW-TaC-CrZr

Total 69.52

* The corresponding alloy of Series IV is given in parenthesis.
** Positive sign refers to an increase in weight percent of the element.